THE INSTITUTION

OF

MECHANICAL ENGINEERS.

ELEVENTH REPORT TO THE
ALLOYS RESEARCH COMMITTEE:

ON

SOME ALLOYS OF ALUMINIUM
(LIGHT ALLOYS).

BY

WALTER ROSENHAIN, B.A., D.Sc., F.R.S.,
SYDNEY L. ARCHBUTT, F.I.C., AND
D. HANSON, D.Sc.,

FROM

THE NATIONAL PHYSICAL LABORATORY. TEDDINGTON
WHERE THE RESEARCH WAS CARRIED OUT UNDER THE
AUSPICES OF THE ALLOYS RESEARCH COMMITTEE.

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ELEVENTH REPORT* TO THE ALLOYS RESEARCH COMMITTEE:

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By WALTER ROSENHAIN, B.A., D.Sc., F.R.S.,
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TABLE OF CONTENTS.

CONTENTS.

Section I.

Introduction. The present Report to the Alloys Research Committee differs in many ways from all its predecessors. In the first place, a period of approximately eight years has elapsed since the publication of the last Report. This long delay has not been due to any cessation or suspension of the work, but has arisen mainly from the fact that under war conditions publication on such a subject was not permissible. Further, at the outbreak of war, the research on the ternary alloys of aluminium with zinc and copper had reached an advanced stage. The period of delay due to war and post-war conditions has thus been superposed upon what would otherwise have been an approximately normal interval between successive Reports.

The length of time thus covered entails the consequence that a very large amount of material has to be treated in the present Report, and this material is very much augmented by the fact that, as a result of the importance attaching to aluminium alloys under war conditions, the research has, during the past five years, been pursued with increased intensity and—as is indicated below—with much larger resources. The mass of material with which the Authors have had to deal is thus very formidable indeed, and any attempt to give an account of it in the fully detailed manner adopted in earlier Reports would have been impossible, owing to the prohibitive bulk of such a Paper. As it is, the Authors have used their utmost endeavour to present the main results of their work in a concise manner, omitting details of experimental methods and even purely numerical data wherever possible, relying mainly upon graphs and diagrams for the presentation of their results. Great care has been given to the preparation of these graphs, and the Authors believe that numerical values of full accuracy can be read from the Figures. A further very necessary abridgment of the present Report has also been made by giving only in the briefest outline the results of the more theoretical portions of the researches on the structure and constitution of the alloys, leaving more detailed publication of this purely scientific work to be made elsewhere by those of the Authors

and their colleagues and assistants directly concerned with that section of the work. It is hoped, however, that Section X, dealing with this branch of the work, is sufficiently full to indicate the more immediate practical bearings of the most scientific part of the research.

The very exceptional conditions in which the researches described in the present Report have been conducted have affected them in several important ways. After the publication of the Tenth Report to the Alloys Research Committee in 1912 attention was concentrated on the systematic study and exploration of the ternary alloys of aluminium with zinc and copper. This portion was approaching completion when the outbreak of war gave a new direction and a new impetus to the work. The steady, systematic exploration of a single group or system of alloys, intended ultimately to furnish the foundation for a future industry, was no longer adequate. The practical application of aluminium alloys, which had formerly been very limited, sprang at a bound into national importance, and results capable of immediate practical application were urgently needed. Fortunately, the material already accumulated in the earlier stages of the research proved in part directly useful and, as a whole, served as a basis for the wide-sweeping exploration of a large range of alloys which was then begun. As a consequence of this development the present Report deals with a wide range of alloys instead of with an exhaustive research on any particular group, with the exception of the copper-zinc-aluminium alloys which had been studied prior to 1914. Much of the data contained in the present Report must therefore be regarded as of a more or less exploratory nature, although it will be seen that in regard to certain alloys of special importance or promise, very detailed investigations have been made.

Another consequence of war conditions and of the increased recognition of the importance of aluminium alloys has been that the Authors were enabled to pursue their work on this subject with the aid of very much greater resources than had formerly been the case. While earlier Alloys Research Reports record researches which were the work of two or three individual investigators, the present Report

contains the results of an organized research carried out, practically, by the whole staff of the Metallurgy Department of the National Physical Laboratory, assisted, in regard to many special matters, by their colleagues in other Departments of the Laboratory, including particularly Engineering (mechanical tests, etc), Metrology (measurements), and Heat (thermal conductivities), The fact that the research has been carried out by the united and co-ordinated efforts of so large a number of workers (more than thirty individuals have contributed in varying degrees) adds very much to the difficulty of the present Authors in presenting the Report, particularly from the point of view of authorship. In doing it in their own names, therefore, they wish to record the fact that they do so mainly in their capacity as senior meml rs of the Staff who carried out the work. In individual acknowledgment they wish to mention particularly the following names : -

Mr. S. A. E. Wells for his important help in the work of the foundry and rolling-mill and in assisting to translate the work of the Laboratory to full industrial scale at the Cardington Works (now the Royal Airship Works).

Messrs. C. B. Marryatt, B.Sc., F. Tritton, and J. D. Grogan, B.A., for their skilled assistance in various portions of the research.

Miss M. L. V. Gayler, M.Sc., and Miss P. M. Routh, B.Sc., for their important contributions to Section V., on the constitution of the alloys and their hardening.

Dr. J. L. Haughton and Miss K. Bingham, M.Sc., for contributions in regard to the constitution and structure of alloys of aluminium containing copper and nickel.

Mr. W. H. Withey, B.A., and Mr. P. G. Ward, B.Sc., for the very large amount of difficult work on the chemical analysis of more than 2,000 samples of metals and alloys ; and

Mr. G. Glaysher, for his great skill and patience in the constant maintenance, and in the construction, of numerous mechanical appliances used throughout the work.

Beyond the Metallurgy Department of the Laboratory, the Authors' thanks and acknowledgments are due particularly to Dr.

The Authors were fortunate in being kept in the closest touch with the whole of this work through the activities of the Light Alloys Sub-Committee of the Advisory Committee for Aeronautics. They were thus enabled to exchange ideas, data, and specimens with other workers, including particularly Professor F. C. Lea, of Birmingham, and Professor C. A. Edwards, at that time associated with the Royal Aircraft Establishment at Farnborough, and later with Mr. F. W. Dyson of that establishment.

To the Light Alloys Sub-Committee, under the chairmanship of Sir Henry Fowler, K.B.E., and later of Professor (then Commander) C. F. Jenkin, C.B.E., also, the Authors' results, particularly those relating to castings, were communicated from time to time as they were obtained, and thus rendered immediately available for national purposes. At the same time, the Authors' conclusions were thus submitted to the most searching of practical tests. The data communicated to and in part published, or in course of publication, by this Committee have necessarily been largely drawn upon in the preparation of the present Report. A very large amount of additional work has, however, been incorporated in it.

In addition to the larger staff and wider range of outside co-operation which became available as a result of war conditions, the Authors were also enabled to carry on their work on a larger and more satisfactory scale by additional resources in plant and equipment placed at their disposal. This was partly the result of the support given to their work by the Advisory Committee for Aeronautics (under the Presidency of the late Lord Rayleigh) and partly by additional funds made available through the efforts of the Alloys Research Committee of the Institution of Mechanical Engineers, who not only themselves provided a larger grant, but also secured increased financial support from the Department of Scientific and Industrial Research. The latter contribution resulted in the provision of a much larger and better-equipped experimental foundry, part of which is illustrated, in connexion with Section II of this Report, in Fig. 45, Plate 5. Perhaps the most important aid to the progress of the present research, however, is constituted by the experimental rolling-mill, which is briefly described and

Illustrated in Section III (page 69) and Fig. 17, Plate 6. This experimental mill was installed *prior to the war*. In regard to its design and general arrangement the Authors are indebted to the makers of the plant, Messrs. Robertson, of Bedford, and also for much advice and assistance of the most valuable kind to Mr. F. Tomlinson, of the Broughton Copper Company, Manchester.

A further word of explanation is perhaps required in view of the fact that publication of the present Report has been delayed for two years after the cessation of hostilities. Apart from the inevitable delays associated with the difficulties of post-war conditions, this has arisen from the circumstances in which much of the work had been done. Under the pressure of war, efforts had been directed to the immediate attainment of practical results, and on reviewing the work done it became evident that from the point of view of a satisfactory study of some of the problems, large gaps remained to be filled. The range of the work has been so wide that it has not proved possible, in any reasonable time, to fill, or even approximately to fill, many of these gaps, but an effort has been made to secure such data as would lend to the work the greatest amount of value, both from a practical and a scientific point of view, which could be attained in the circumstances. This work has continued up to the moment of writing, as many omissions could only be realized when the data were marshalled for publication. The Authors are, therefore, well aware that there are numerous points in the Report where additional data are obviously desirable, but they have not thought that these gaps were of sufficient importance to justify further delay in publication.

With regard to the general results embodied in the Report, it is almost impossible to offer any brief summary that could be more than a catalogue of facts and figures, and all that can be attempted here is to draw attention to a few of the more striking results. In regard to casting alloys, the systematic study of the ternary alloys of copper, zinc and aluminium has led to the recognition of the valuable properties of these for casting purposes where the castings are not exposed to high temperatures. On the other hand, however, the most important result established in the Report in regard to

castings is probably that relating to the great strength obtainable by means of heat-treatment in castings of the alloy " Y " (containing 4 per cent copper, 2 per cent nickel and $1\frac{1}{2}$ per cent magnesium). Tensile strengths as high as 21 tons per square inch have been recorded by the Authors, and equally high values have been recorded by the R.A.W. For use at high temperatures also, such as those occurring in aeroplane engine pistons, castings of this alloy show a very marked superiority over those of any other aluminium alloy yet available, and for this purpose also the alloy is materially improved by heat treatment. In regard to wrought alloys, the Report shows remarkable results obtainable in the first place by a simple ternary copper-zinc-aluminium alloy, called alloy "A" or "3/20" in recognition of its content of 3 per cent of copper and 20 per cent of zinc. This alloy, which is cheap and comparatively easily produced, attains a tensile strength of 27 tons per square inch with an elongation of 18 per cent on 2 inches. More striking are the results shown by alloys containing manganese and magnesium in addition to copper and zinc. Here (alloy " E ") tensile strengths exceeding 40 tons per square inch are recorded, with elongations of about 12 per cent on 2 inches. On the other hand, alloy "Y," in the rolled or heat treated condition, attains strengths of 26 tons per square inch with elongations of 20 per cent on 2 inches, thus giving results approximately equal to those obtained with alloy " A " and " Duralumin," but with the important advantage that it appears to offer greater resistance to corrosion than other aluminium alloys, while its power of retaining its strength when heated gives it appreciable advantage, particularly in regard to fatigue at slightly elevated temperatures.

One of the most important results achieved by the present research has been the successful working, by forging, extrusion, rolling, spinning, stamping, etc., of these alloys, not only on account of their remarkable properties but because they represent important types which were formerly regarded as unworkable. Some examples of the forms in which these materials have been successfully produced are illustrated in Figs. 1 and 2, Plate 1.

In addition to their strength properties, the alloys have also

been studied from the point of view of stability and permanence.
The constancy of dimensions of a large series of copper-zinc-
aluminium castings has been demonstrated over a period of more
than seven years, but on the other hand it has been discovered
that these alloys, both in the cast and in the wrought state, undergo
a gradual ageing process which is shown to continue for several
years at least. This process tends to render the alloys slightly stronger
and harder, but no sign of deterioration or embrittling has been
noted. In the case of " Duralumin " also, a slight further hardening
has been observed in material kept in store for nearly ten years.
The nature of the changes which occur in alloys containing
magnesium when they undergo hardening by ageing after quenching
from a high temperature has been studied and, it is believed,
elucidated. The results throw a considerable amount of new light
on the processes involved in hardening and tempering generally,
and also serve, to a considerable extent, to explain the general
nature of the processes involved in the gradual " ageing " of some
of the alloys.

For convenience of reference and arrangement, the present
Report has been subdivided into five main Sections and various
sub-sections as follows :—

Section I.—Introduction.

Section II. Cast Alloys : --

    (*a*) Alloys for use at ordinary temperatures.
    (*b*) Alloys for use at high temperatures.
    (*c*) Heat-treatment of castings.

Section III.—Wrought Alloys :—

    (*a*) Ternary Alloys of Aluminium with Zinc and Copper.
    (*b*) Development of Alloy " A " (" 3/20 ").
    (*c*) Development of Alloys " E," " F," " G," and " Y."
    (*d*) Extrusion of Aluminium Alloys.
    (*e*) Summary and Comparison of the Properties of Wrought
        Alloys.

Section IV.—Permanence of the Alloys :—

    (*a*) Ageing.

    (*b*) Stability of Dimensions.

    (*c*) Fracture under Prolonged Loading (" Season Cracking ").

    (*d*) Corrosion.

Section V. —Constitution of the Alloys :

    (*a*) The Copper-Zinc-Aluminium System.

    (*b*) The Silicon-Iron-Aluminium Sytsem.

    (*c*) The Magnesium-Silicon-Aluminium System.

    (*d*) The Hardening of Alloys containing Magnesium and Silicon.

SECTION II.

CAST ALLOYS.

The present section deals with aluminium alloys from the point of view of their use and properties as castings. In many cases the same alloys are dealt with in regard to constitution in Section V (page 196) and in regard to their behaviour on forging and rolling etc., in Section III (page 41). The ageing and stability properties of the same alloys are dealt with in Section IV (page 155).

II(*a*). *Alloys for Use at Atmospheric Temperature.*

In continuation of the researches begun in connexion with the Tenth Report to the Alloys Research Committee, the earlier stages of the present research were directed to the fuller exploration of the properties of the alloys of copper, zinc and aluminium. A considerable number both of exploratory heats and of larger meltings were made and the resulting castings, both sand and chill, tested in the usual manner. The results of these tests are indicated in the diagram Fig. 3, in which the compositions of the various alloys are plotted on the base of a portion of the usual equilateral triangle, and the data for ultimate stress inserted as numbers at the

FIG. 3.—*Sand Castings.*
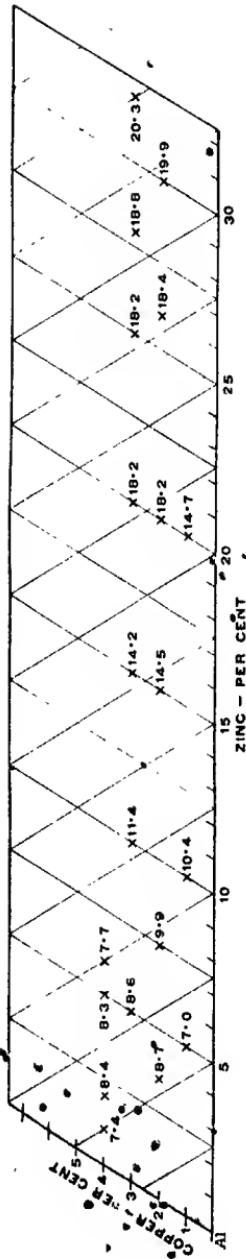
FIG. 4.—*Chill Castings.*
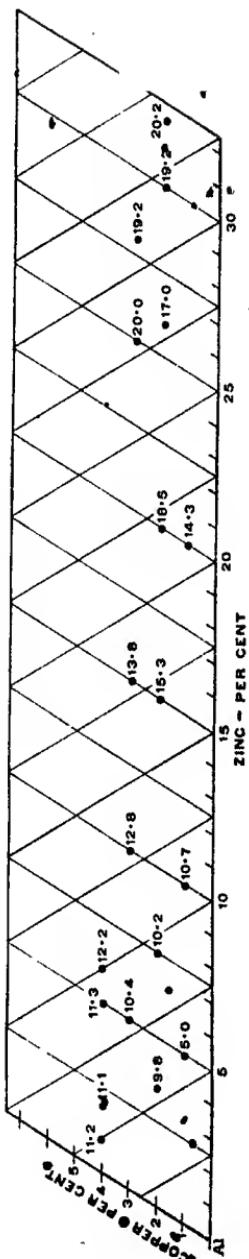
corresponding points of the diagram. Fig. 3 relates to sand castings, and Fig. 4 to chill castings. In regard to the sand castings, it will be seen that tensile strength as high as 19 tons per square inch has been obtained, but this is found in alloys containing about 28 per cent of zinc, and these are comparatively heavy. Greater interest and importance, therefore, probably attaches to alloys containing in the neighbourhood of 15 per cent of zinc, and in

FIG. 5. —*Tensile Tests.*

(Numbers on Graphs indicate Copper Content.)
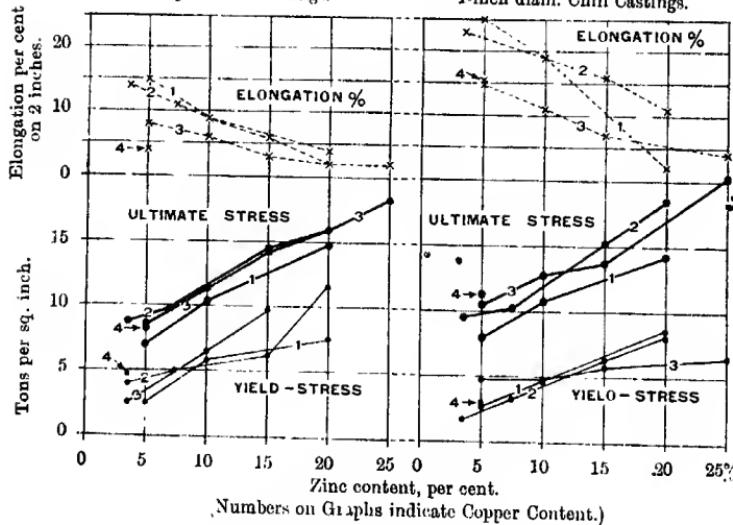
these tensile strengths of the order of 14 tons per square inch are recorded. These data, however, should be read in connexion with what is said in Section IV (*a*) (page 155), concerning the ageing properties of these alloys.

The same results can be shown in a different way by plotting ultimate stress data against zinc content for a series of alloys having a constant copper content. This has been done in regard to sand and chill castings in Fig. 5. It will be seen that in each series the tensile strength rises steadily with increasing zinc content, but that the addition of copper makes very little difference as between 2 per

cent and 3 per cent of copper, where the two graphs are almost exactly superposed.

The data concerning the tensile strength of ternary alloys with copper, zinc and aluminium as indicated in Figs. 3, 4 and 5, which have just been discussed, suggest at first that the higher zinc content, with corresponding considerably higher tensile strength, gives the advantage to the alloys containing from 20 to 25 per cent of zinc. For practical purposes, however, there appear to be certain objections to these alloys, arising mainly from the fact that they are very weak and tender when hot and are therefore liable to accidental injury in the foundry. While difficulties of this kind could no doubt be overcome by more careful manipulation, it appears doubtful whether the use of these stronger and at the same time considerably heavier alloys is desirable. For many purposes, the thickness of a casting cannot be reduced below a certain value, and even if the metal does not possess the relatively high tensile strength of these denser alloys, the minimum thickness which can be used yet leaves it strong enough for practical requirements. If this is the case, then the lower density of the alloys containing rather smaller proportions of zinc is a decided advantage. For this reason the alloys which have been selected for most general use for castings not exposed to high temperatures are those containing rather less than 15 per cent of zinc. Typical of these is the alloy now widely known as " L5 " on account of the number of the official specification in which it has been described. This alloy contains from $12\frac{1}{2}$ to $14\frac{1}{2}$ per cent of zinc, and from $2\frac{1}{2}$ to 3 per cent of copper. A freshly cast chill test-bar is required to show a tensile strength of not less than 11 tons per square inch and an elongation of 4 per cent on 2 inches. In view of the data given above, it will be seen that these requirements are extremely moderate and can readily be exceeded. The actual strength of castings made of this material after a few months' ageing will, however, be very considerably greater than that indicated by the results of tests made on freshly prepared test-pieces. This particular casting alloy has been used on a very large scale, and has on the whole proved very satisfactory. Results recently obtained with

heat-treated castings of an entirely different composition, however, suggest that for many purposes castings of the copper-zinc-aluminium type are likely to be superseded by those of alloys capable of improvement by heat-treatment.

Density.

The density at atmospheric temperature of the alloys studied, in both the cast and wrought conditions, has been determined by the Metrology Department of the Laboratory. Results for the wrought alloys are given in Sections III (a and c) (pages 45 and 142). The results for the cast alloys of the ternary copper-zinc-aluminium series containing 1, 2 and 3 per cent copper, and for the alloy containing copper 4, and zinc 5 per cent respectively, are shown in the graphs of Fig. 6. So little difference has been found between sand and chill cast specimens that in the case of the latter condition the graph of the 3 per cent copper series only is included for comparison.

Microstructure of the Alloys.

The microstructures of the alloys of the ternary group, copper-zinc-aluminium, both in the sand and chill cast condition, have been systematically examined. Of the large number of microstructures which have been recorded, it is only necessary to reproduce here typical examples, relating to the sand and chill cast material of alloys containing respectively 1 per cent of copper with 5 per cent and 20 per cent of zinc, and of those containing 3 per cent of copper with the same proportions of zinc. These are shown in Figs. 7 to 10 Plate 2, all under a magnification of 150 diameters. Figs. 7 and 8 are typical of the cast alloys containing 3 per cent copper in the sand and chill cast state respectively, and Figs. 9 and 10 of those containing 1 per cent copper. In these figures it will be seen that the presence of the aluminium-copper compound $CuAl_2$ makes itself evident. In addition there are the usual signs of coring in the solid solution of zinc in aluminium, so many typical examples of which have been illustrated in the Tenth Report to the Alloys Research Committee.

Comparison of these photo-micrographs, representative of alloys of this type in the cast condition, with those seen in Figs. 41 to 44 (Plate 4), relating to alloys of the same type in the rolled condition, is instructive as showing the influence which repeated

Fig. 6. – *Density.*

Grammes per cubic centimetre.
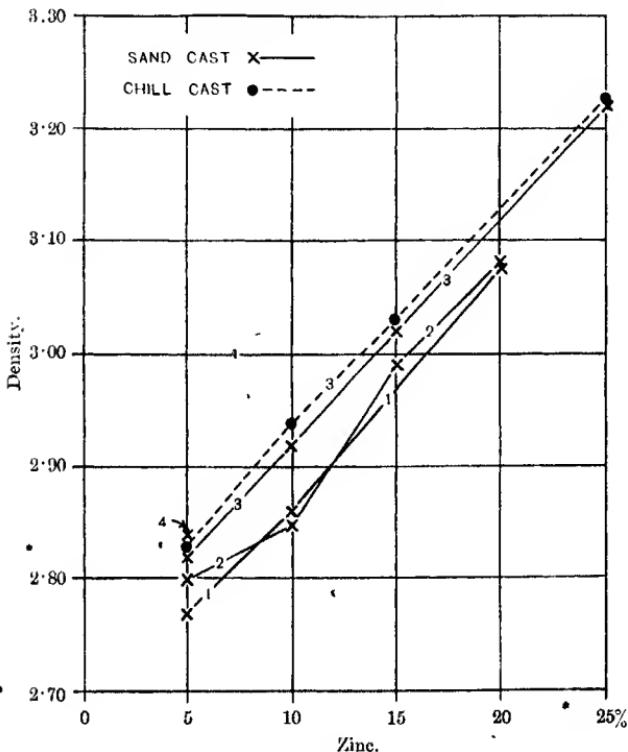
heating and mechanical working exerts upon the microstructure of these materials.

II(b). *Cast Alloys for Use at High Temperatures.*

The advantages to be gained from the use of aluminium alloys

as pistons in aeroplane and other high duty internal combustion
engines having been demonstrated in practical trial as well as
indicated by theoretical consideration, the problem of finding the
most suitable casting alloy for this purpose became of very

FIG. 11.– *Tensile Tests at High Temperatures* (*Chill Castings*).

No. 36—Alloy " Y " (Cu 4%, Ni 2%, Mg 1·5%).

14/1   Alloy containing Cu 14%, Mn 1°.
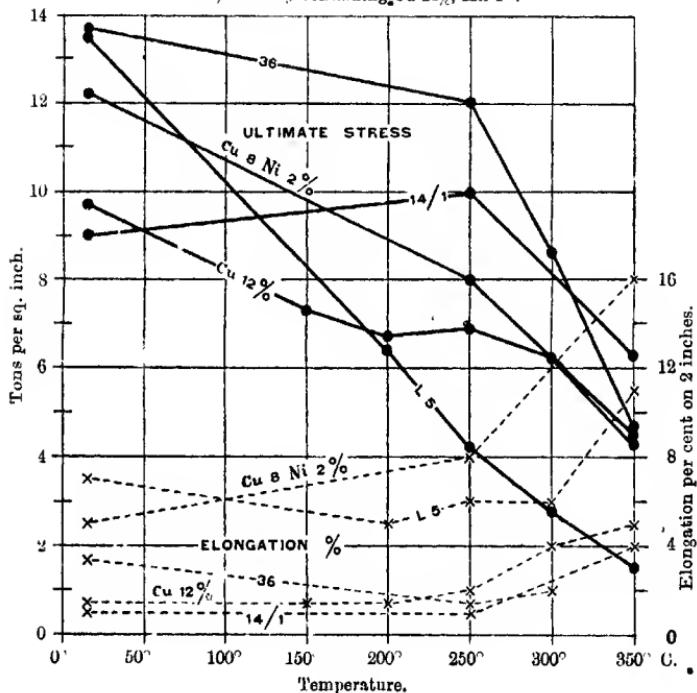
considerable importance.   In the course of the present research
the Authors have devoted a very large amount of attention to this
subject.

Reference has already been made to the fact that alloys of
aluminium containing considerable proportions of zinc, even in
the presence of a few per cent of copper, are very weak when hot.

Since the behaviour of light alloys at high temperatures has become
a matter of considerable importance in connexion with their use
in aeronautical engines, mainly as pistons, their strength at high
temperatures has received a considerable amount of attention
in the course of the present research. Results of tests at high
temperatures on the zinc-aluminium binary alloys have already
been described in the Tenth Report to the Alloys Research Committee.
These show that the tensile strength of such alloys falls off rapidly.

Further tests at high temperatures have shown that the ternary
alloys of copper, zinc and aluminium behave in a very similar
manner. The results of one series of tests of this kind, relating to
the alloy " L5," are included in the graphs of Fig. 11 which represent
the behaviour of a number of typical alloys at high temperatures.
There it will be seen that, starting with a tensile strength of about
13·5 tons per square inch, this falls off at 250° C. to little more
than 4 tons per square inch, and at 350° C. drops to about 1·5 tons
per square inch. This very rapid fall in strength with rising
temperature has led to the abandonment of these alloys for uses
where high temperatures are involved, although at the temperature
200° C., which appears to be of the greatest importance in connexion
with aeroplane engine pistons, the alloy " L5 " has approximately
the same strength as the alloy containing 12 per cent of copper
(remainder aluminium), which has been very extensively used for
that very purpose.

Some guidance as to the behaviour of other alloys at high
temperatures had already been obtained at the time when
this problem arose, from experience of the behaviour of various
alloys in the extrusion press (see Section III (d) below). It has
been found, for instance, that alloys of aluminium and copper
containing more than about 4 per cent of copper prove very stiff and
intractable for extrusion, even at high temperatures, and it was
thought, therefore, that these alloys might be found to preserve their
strength at high temperatures better than others. A sample of
such an alloy, of which an aero engine piston had actually been
made, had also been submitted to the Authors for test by what was
then the Technical Section of the Royal Naval Air Service. An

extensive series of tensile tests at high temperatures was therefore
undertaken and the properties of a large number of alloys at high
temperatures were explored. The results obtained with the more
important of these are indicated in the graphs of Figs. 12 to 19.
The only data available from such tests, as carried out by the

Fig. 12.—*Chill Castings.*
6%, 8%, 12% and 14% Copper.
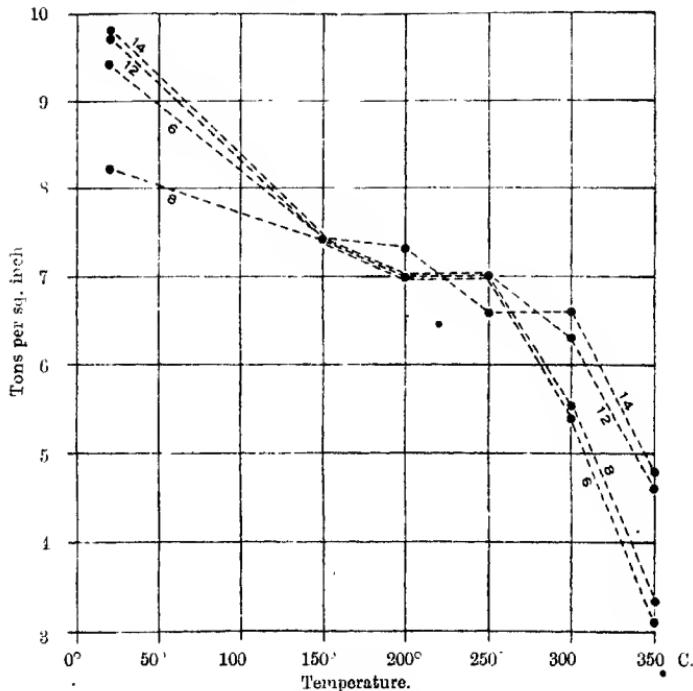
Authors, were the ultimate stress and elongation. No attempt to
measure a yield-point was made, although subsequently Professor
Lea succeeded in determining the elastic limit of these materials at
high temperatures by means of a special extensometer which was,
however, unsuited for use in the type of testing-machine available
to the Authors. The data represented by Fig. 12 relate to

alloys of copper with aluminium containing 6, 8, 12, and 14 per cent
of copper respectively.   It will be seen that in the neighbourhood
of 200° to 250° C. all the alloys of aluminium with copper, have
practically the same tensile strength, but that at somewhat
higher temperatures, up to 350° C. the alloys containing 12 per
cent and 14 per cent of copper are somewhat better than those
containing lower proportions.   The alloy containing 12 per
cent of copper, generally known by its Air Board Specification
Number " L8," has been so extensively used, that it has been
adopted as the standard for comparison throughout the remaining
figures relating to high temperature tensile tests of cast aluminium
alloys in the present Report.

The first effort to improve upon the simple aluminium-copper
alloys for use at high temperatures, was made by the addition to
them of a small amount of manganese.   The influence of manganese
in stiffening materials at high temperature had previously come
under the Authors' notice, and the results in the present instance
justified their anticipation.   The influence of manganese on these
alloys is clearly shown in the graph, Fig. 13, where results of
high temperature tensile tests on alloys containing 4 per cent and
14 per cent of copper plus 1 per cent of manganese in both the sand
and chill cast condition are shown compared with the properties of
the 12 per cent copper alloy chill cast.   It will be seen that in every
case the alloys containing 1 per cent of manganese show an actual
increase in tensile strength up to temperatures of 250° C.   After
that temperature has been passed, their strength falls off at much
the same rate as that of the pure copper alloys, but owing to the
initial rise, their strength at 350° C. is still very much better than
that of the pure copper alloys.

This effect of manganese appeared to be so striking that it was
investigated more closely, with a view to ascertaining whether the
addition of greater or smaller amounts of manganese might produce
a better result than that already obtained by the addition of 1 per
cent.   Rather curiously, it was found that the addition of 1 per cent
gave by far the best results.   This is shown in Fig. 14, where
it is seen that in a graph representing the variation of tensile strength

at various temperatures with different amounts of manganese, there is a distinctly marked maximum about 1 per cent of manganese.

So far as tensile tests at high temperatures are concerned, it

FIG. 13.

1 = Copper 14 , Manganese 1 , Chill Castings.
2 =   „   „   „   „   Sand   „
3 =   „   8   „   „   Chill   „
4 =   „   „   „   „   Sand   „
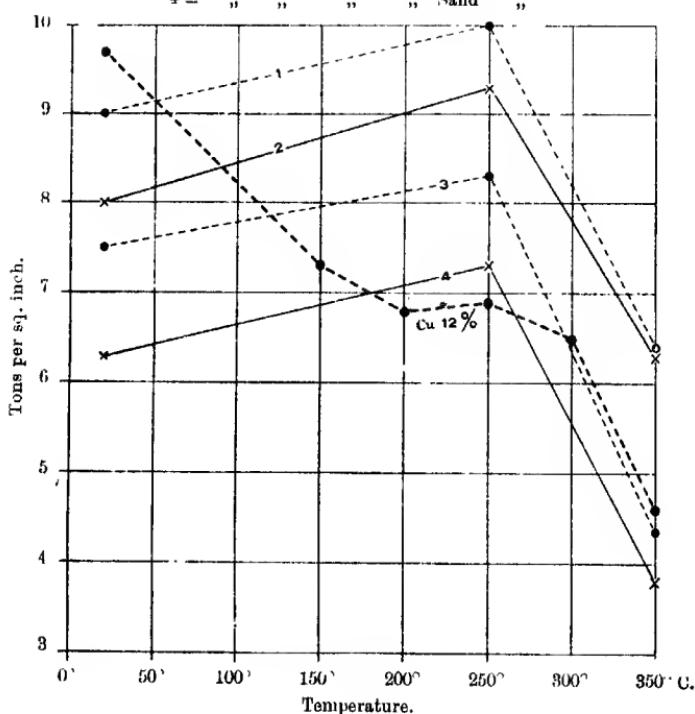
might have been thought at this stage that the problem of finding a light alloy suitable for use at high temperatures, and particularly in aeroplane engine pistons, had been satisfactorily solved. Other properties besides tensile strength at high temperatures, however, influence the choice of an alloy for this purpose. In the case of

the alloy containing 11 per cent copper, 1 per cent manganese, for instance, it was found that it could not at first be so readily produced in the aluminium foundries as other alloys. While there can be no doubt that difficulties of this kind could have been successfully surmounted, and in fact in one or two cases they were successfully overcome, yet under war conditions it appeared undesirable to introduce for extensive practical purposes a material which, at the

FIG. 14.—*Sand Cast Bars.* 14 *Copper.*

Effect of varying Manganese content.
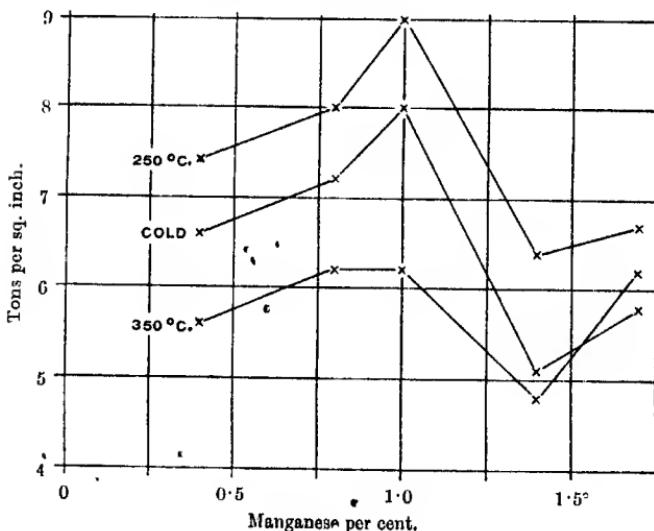
outset, at any rate, might present difficulties of manufacture. It was further thought for a time that this alloy might be inferior to others in regard to thermal conductivity, and for these combined reasons the search for other and, if possible, superior alloys suitable for use in pistons and for other purposes at high temperatures, was continued. A considerable number of miscellaneous alloys were explored. These included alloys with iron, molybdenum, tungsten, chromium, vanadium, and others. The properties at high temperatures of some of these are illustrated in the graphs of

Fig. 15. It will be seen at once, on reference to this Fig. that none of these offer any advantage whatever as compared with the alloy of copper, manganese, and aluminium, described above.

FIG. 15. *Chill Castings.*
1 = Copper 12 , Manganese 1 , Chromium 1 .
2 =   „   3 , Cobalt 3 .
3 =   „   14 , Vanadium 0·7 .
4 =   „   „   Iron 1 .
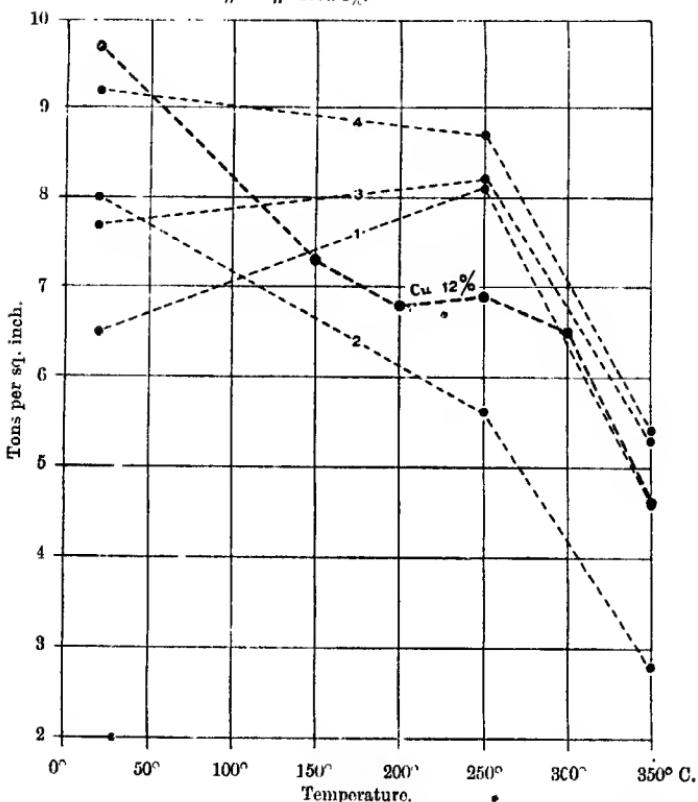
In the majority of these alloys it was found that a comparatively small addition of the second element renders the material extremely brittle and unsatisfactory from a casting point of view. It had

been hoped that, in these alloys the presence of an inter-metallic
compound having a very high melting-point and a considerable
degree of harshness, might render the material stiff and hard at
high temperatures.   Unfortunately, it was found that these high-
melting inter-metallic compounds produce in the alloys, during
casting, a rigid skeleton from which the remaining metal, which is
still liquid at a temperature when the skeleton has solidified, tends
to drain away, leaving an unsound porous casting.

In view of the results just quoted, attention was next
concentrated upon determining the influence of iron and nickel
upon the more usual aluminium alloys.   A series of preliminary
experiments convinced the Authors that the addition of iron to
copper-aluminium alloys, although it did not produce any marked
deterioration in the mechanical properties, did not tend to produce
any very important improvement in their behaviour at high
temperatures, while at the same time it undoubtedly increased the
difficulties in the foundry.   Other workers have pursued this
particular matter very much further, and their results on the whole
confirm the view arrived at by the present Authors at an early
stage of the work.   The study of the effect of the addition of nickel,
together with further additions either of manganese or of magnesium,
has led, on the other hand, to results of first-rate importance.   One
of the alloys, which has been distinguished throughout the present
Report by the letter " Y," which resulted from these investigations
for the purpose of producing piston alloys, has proved extremely
useful in other directions, and in the form of rolled bars and sheet
and similar products promises to be of very considerable importance
(see Section III (c) page 111).

The first step in investigating the influence of nickel on alloys
of this kind consisted in a series of tests made upon alloys containing
from 8 to 12 per cent of copper and 1 to 3 per cent of nickel.   The
results of tensile tests at high temperatures on typical examples of
this series are shown in the graphs of Fig. 16.   For purposes
of comparison the graph for the alloy containing 12 per cent of
copper alone is included in this Fig. Comparison of these graphs
shows at once the beneficial effect of nickel, but there is still a

distinct fall in tensile strength between the ordinary temperature
and 250° C., followed by a much more rapid fall at still higher

FIG 16.—*Chill Castings.*

1 = Copper 12%, Nickel 3%.
2 =   ,,   ,,   ,,   2%.
3 =   ,,   9%,   ..   ,, .
4 =   ,,   8%,   ,,   1%.
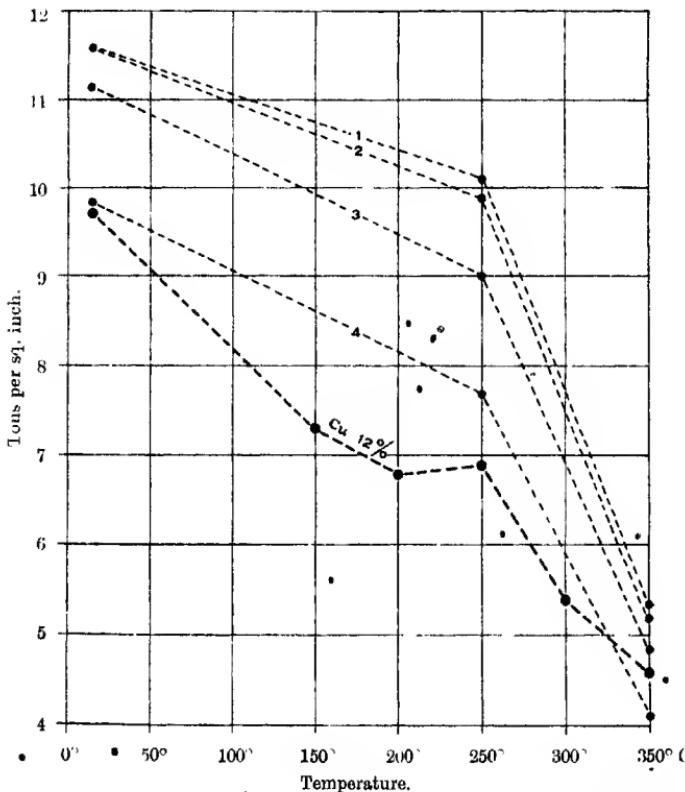
temperatures.   The next step consisted in testing the effect of the
addition of magnesium to alloys containing both copper and nickel.
In the first instance the alloy containing 8 per cent of copper and

FIG. 17.— *Chill Castings.*

33A — Copper 8%, Nickel 2%, Magnesium 1 %.
34 — ,, 8', ,, 2,⁰, ,, 1·5%.
33 — ,, 8', ,, 2, ,, 2 %.
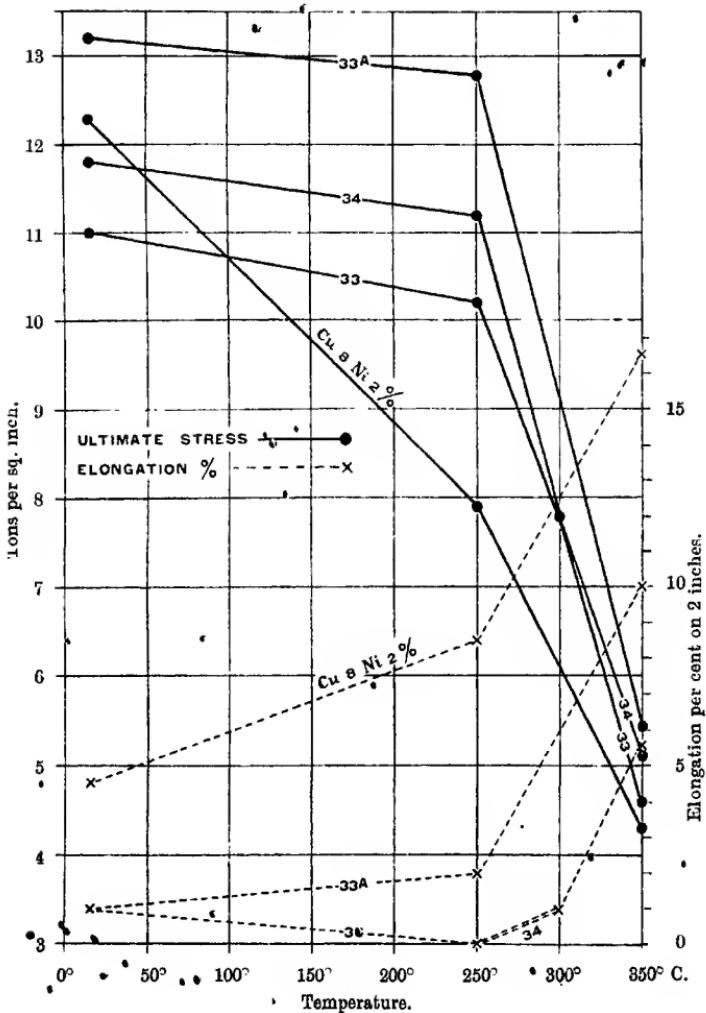
2 per cent of nickel was chosen for this purpose. The alloys tested were made to contain as nearly as possible 8 per cent of copper, 2 per cent of nickel, and 1, 1½ and 2 per cent of magnesium,

FIG. 18.—*Sand Castings.*

36—Alloy " Y " (Cu 4 , Ni 2 , Mg 1·5).

34—Alloy containing Cu 8 , Ni 2 , Mg 1·5 .
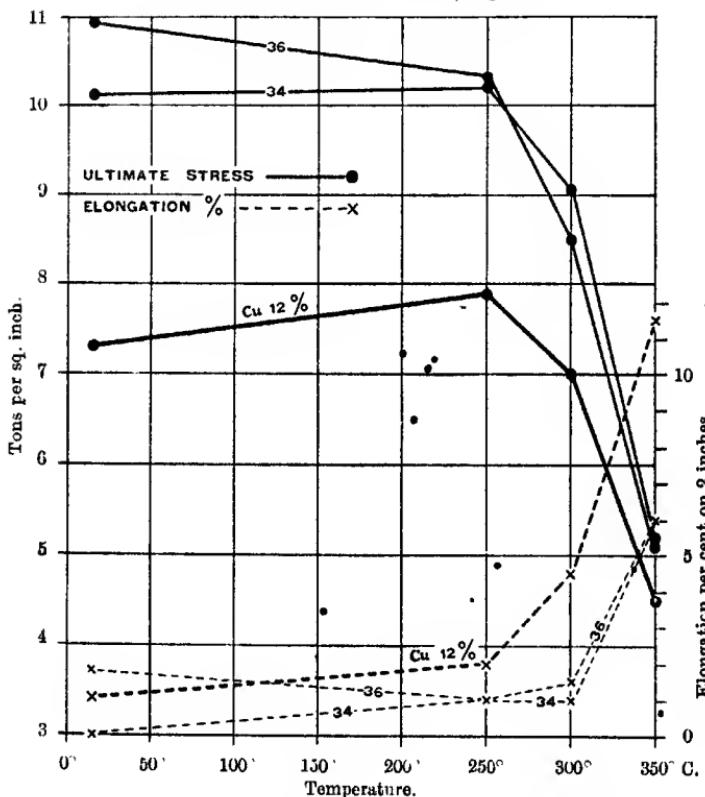
respectively. Tests were made on both sand and chill castings, but for purposes of comparison it is sufficient to confine ourselves in this case also to the tests on chill castings. These are indicated in the graphs of Fig. 17. Comparison of the graphs of this

Fig. shows that the addition of magnesium has produced an important improvement in the alloys. Particularly the alloy containing only 1 per cent of magnesium at a temperature of 250° C. shows a tensile strength of close on 13 tons per square inch as compared with 8 tons per square inch in the alloy having the same copper and nickel content in the absence of magnesium, the alloy with copper alone being still weaker at that temperature.

No further details need be given of the intermediate stages of the investigation, in view of the fact that it was shortly afterwards discovered that in the presence of both nickel and magnesium the copper content of these alloys could be safely reduced, with considerable advantage to the lightness of the alloy, and also to its thermal conductivity.

The results of high temperature tensile tests on both sand and chill cast test-pieces of alloy " Y " (4/2/1½) are shown in Figs. 18 and 19 relating to sand and chill castings respectively. On both these figures, the best of the alloys containing larger proportions of copper with nickel and magnesium (alloy 34 containing 8 per cent copper, 2 per cent nickel, 1 per cent magnesium) is given for comparison, as well as a graph representing the properties of the 12 per cent copper alloy. It will be seen that in both sand and chill castings the alloy " Y " is slightly superior to the alloy 34, and of course, enormously superior to the alloy containing 12 per cent of copper alone. An interesting and important feature in both sand and chill cast test-pieces of this material is the fact that they show an appreciable amount of elongation, and at really high temperature, above 300° C., this elongation increases very rapidly. It was this observation which led the authors to try forging and rolling of this material, with the successful results described in Section III (c), (page 111).

In order to ascertain the limits of composition within which variation might be allowed without materially impairing the important properties of alloy " Y," a series of tests were carried out with alloys in which the composition had been varied from 1 per cent of nickel to 2·5 per cent, and magnesium from 0·5 per cent to 2 per cent. The results show that small variations of composition

FIG. 19.—*Chill Castings.*

36—Alloy " Y " (Cu 4 , Ni 2 , Mg 1·5).
34—Alloy containing Cu 8 , Ni 2 , Mg 1·5 .
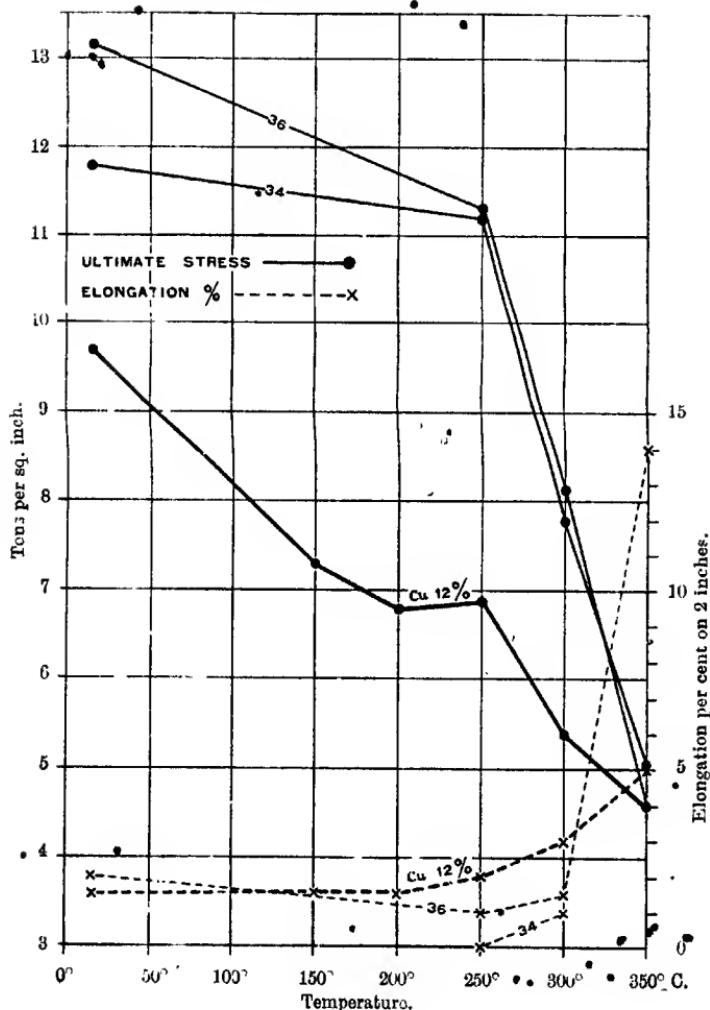
do not appreciably affect the alloy, but that the composition originally
arrived at, namely, 2 per cent of nickel and 1·5 per cent of magnesium
represents the best of the materials tested.

Other Properties of the Alloys.

Apart from tensile strength at high temperatures, which has
been regarded as a fundamental measure of the value of an alloy
for such uses as pistons in internal combustion engines, a number of
other properties are of considerable importance. The question
of the general facility with which an alloy can be cast into more or
less complicated shape, and the facility with which sound castings
can be obtained, are of very considerable importance. Many of the
casting alloys exhibit a marked evolution of gas at a temperature
at or just above the commencement of freezing. Alloy "Y" is certainly
not free from this trouble. There is also in all these alloys a
considerable tendency towards the defects known as "drawing,"
which occurs if the supply of liquid metal to the casting is cut off
before the whole of the casting has solidified, so that contraction
during the final stages of solidification results in local shrinkage.
The avoidance of defects from both these causes is mainly a question
of proper pouring temperature and adequate gating and venting
of the moulds. In the case of alloy "Y," and other alloys containing
magnesium, a special precaution is, however, required. This relates
mainly to the manner of introducing magnesium, which it is found
best to add immediately before the metal is poured into the mould. In
the Authors' experience, it has been found best to add the magnesium
in the form of the pure metal cut into fairly large pieces and to push
these rapidly under the surface of the molten aluminium so that
they can dissolve without coming into contact with the air. This
is readily achieved by the use of a small inverted cup, for which
purpose a salamander crucible through which a number of holes
have been drilled can be satisfactorily used. Such a crucible,
attached to a suitable rod as a handle, is shown in the sketch,
Fig. 20. If magnesium is added in this way, it is easy to
achieve a very close approximation to the desired composition of

the finished alloy.' If magnesium is added at an earlier stage, on the other hand, a variable proportion of that element is liable to be lost during melting, and the alloys are also less likely to be sound and satisfactory in other respects.

With regard to the introduction of nickel to these alloys, some care is also required. Metallic nickel does not readily dissolve in aluminium, and rich alloys of nickel and aluminium become very infusible or, at all events, contain a very infusible body. For this reason, the Authors prefer to prepare a hardener or rich alloy containing only 20 per cent of nickel and to add this in the requisite quantity to molten aluminium when preparing alloys containing nickel. With these precautions, alloys of the "Y" type can be satisfactorily made and cast. In these, and

Fig. 20.
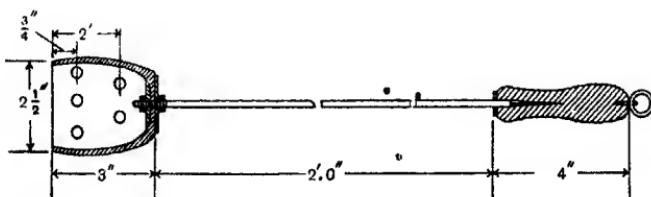
in a great many other light alloy castings, the presence of a considerable number of very minute holes, commonly known as pin-holes, makes itself evident when the surfaces are machined. Although these minute holes give an appearance and suggestion of unsoundness on a very minute scale, actual tests show that castings containing them do not appear to be appreciably weaker than portions of the same casting which are free from this apparent defect. While, therefore, it is undoubtedly desirable to eliminate these minute holes as far as possible, by suitable foundry manipulation, yet their presence need not be regarded as seriously deleterious to the casting.

Casting Contraction.

In order to ascertain whether the differences in the behaviour of various alloys in the foundry could rightly be ascribed to differences

in the amount of contraction undergone by them when cooling
from the liquid state, a series of determinations of the total linear
contraction of a number of alloys was undertaken. Two methods
were employed. In the first of these, a sand mould was prepared
with a pattern 12¾ inches long by ¾ inch square in section. This
was provided with two sharp pointed projections or " pips." 12
inches apart. After drying the mould, and just previous to pouring
in the metal, the distance between the points of the impressions
of these pips in the sand was measured by means of dividers. The
mould was then closed and the casting poured. When sufficiently

FIG. 21.—*Total Linear Contraction of Light Alloys.*

Scale.—Quarter full size.
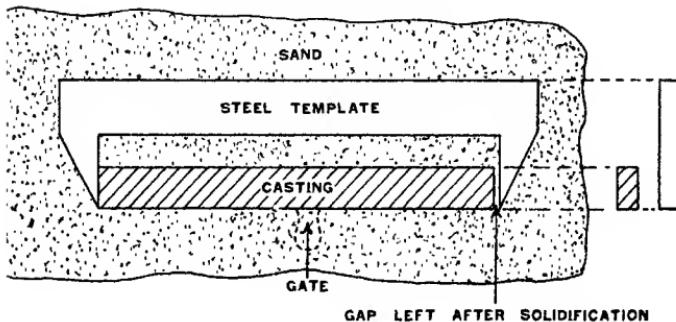
cold, the casting was removed, allowed to cool to atmospheric
temperature, and the distance between the points of the projections
of the castings measured. The difference between these two
measurements gives the contraction. In the second method, a
flat bar 10 inches long by 1 inch wide by ½ inch thick was cast in
sand between the machined faces of a steel template, as shown in
the sketch, Fig. 21. The wooden pattern and the steel template
were moulded together in the sand, and both removed and the
mould thoroughly dried and allowed to cool. The distance between
the faces of the template having been checked, it was replaced in
the mould, the mould closed and the casting poured. Template
and casting were removed from the mould as soon as possible.

When the cast bar had cooled completely, it was replaced between the faces of the template and the small gap due to the contraction of the aluminium casting determined either by the use of feeler gauges or in other ways. The results obtained by both methods are in satisfactory agreement, but finally the method of casting between the faces of the steel template was adopted in preference to the other.

TABLE 1.

Alloy (Percentage Composition).					Contraction per cent (means).
Manganese.	Copper.	Zinc.	Tin.	Aluminium	
—		15	- -	85	1·42
—	2	10	—	88	1·40
...	2·75	13·5		83·75	1·27
—	3	15	—	82	1·25
1	8	—	—	91	1·31
1	14	--	—	85	1·21
—	7	1	1	91	1·19
—	10	-	--	90	1·25
--	12	-		85	1·25
—	10	--	1·25	88·75	1·22
- -	2		1·5	86·5	1·25

The results obtained for a series of eleven alloys are shown in Table 1. It will be seen that the total contractions vary comparatively little between different alloys. The highest contraction (1·40 per cent) is found in the pure binary zinc-aluminium alloy containing 15 per cent of zinc ; on the other hand, the alloy "J5" has a contraction of only 1·27 per cent, while the alloy containing 14 per cent of copper and 1 per cent of manganese has a contraction of 1·21 per cent. In general, the results indicate very clearly that

differences in total contraction between the various alloys are
insufficient to account for the differences in their behaviour described
by those using them in the foundry.   Such difficulties as have
at times arisen in connexion with the casting of certain of the alloys,
therefore, must be ascribed to other causes.

Notched-Bar Impact Tests.

As it was feared that certain of the alloys containing high
proportions of copper and manganese particularly, might prove to
be brittle in practice, a series of notched-bar impact tests both at

TABLE 2.

Alloy.	Energy absorbed in Fracture. (Mean Values). Kg.-Metres.	
	15° C.	250° C.
"L5" . . . • . . .	0·200	0·192
"L11" (Copper 7, Tin 1, Zinc 1 per cent).	0·082	0·015
Copper 9, Tin 2, Zinc 1·5 per cent . .	0·048	0·009
"14/1" (Copper 14, Manganese 1 per cent)	0·016	0·016
Copper 14, Manganese 1, Tin 1 per cent .	0·011	0·007
Copper 6 per cent . . . · . .	0·146	0·097
Copper 8 per cent ·	0·095	0·061
"L8" (Copper 12 per cent) . . ·.	0·033	0·024
Copper 8, Nickel 2, Magnesium 1·5 per cent .	0·013	0·015
"Y" (Copper 4, Nickel 2, Magnesium) 1·5 per cent) . . . . .)	0·042	0·035

ordinary and at elevated temperatures have been carried out by
the Engineering Department of the National Physical Laboratory.
The results of these tests on specimens from chill cast bars, mostly
1 inch diameter, which are tabulated below in Table 2 are not
very conclusive from this point of view.   It is now recognized that
notched-bar impact test values can hardly be regarded as a safe

means of comparing materials differing considerably in character, and the differences between some of the alloys considered in the present Report are wide enough to vitiate any comparison of that kind. For that reason the Authors do not suggest that the relative merits of the various alloys can safely be gauged in any way by comparison of the impact values given in this Table. The only striking fact which appears to emerge from the data obtained, relates to the influence of tin on the impact strength of the alloys at elevated temperatures. Comparison of the figures obtained with alloys differing only in that one of them contains 1 per cent of tin, whereas tin is absent in the others, shows clearly the deleterious influence of tin in this respect.

Thermal Conductivity.

The main advantage derived from the use of aluminium alloy pistons in internal combustion engines, such as those used on aeroplanes, is due, not so much to any small saving in weight even of reciprocating parts which may be effected in this way, as to the very considerable gain in the efficiency of the engine owing to the greater heat-conducting power of the aluminium pistons. In the case of air-cooled engines, the use of light alloys having high thermal conductivity avoids the distortion arising from unequal cooling of the leading and trailing sides of the cylinders. Even in water-cooled engines, however, the higher thermal conductivity of the aluminium piston allows the use of considerably increased compression ratios, which lead to an increase of the power of the engine, and at the same time to a decrease of petrol consumption per horse-power-hour. In view of the importance which thus attaches to thermal conductivity, a series of careful determinations of this property have been made by Dr. E. H. Griffiths, of the Heat Department of the National Physical Laboratory. The methods employed in these measurements have been described elsewhere, and here it is only necessary to quote the results. These are embodied in Table 3.

It will be seen from this Table, that the thermal conductivities of the various alloys do not differ to any very large extent,

TABLE 3.
Thermal Conductivity.

Thermal Conductivity.

Composition by Analysis.	Sand.				Chill.			
	As cast.		Annealed 450°.		As cast.		Annealed 450°.	
	100°.	300°.	100°.	300°.	100°.	300°.	100°.	300°.
Copper (1). 12*	0·34	0·37	0·36	0·38				
Tin. Zinc. Copper (2). 0·82 0·88 6·64	—	—	– –	—	0·39	0·41	0·38	0·40
Copper. Zinc (3). 2·75* 13·5*.	0·32	0·36	—	—	0·31	0·37	0·34	0·35
Silver. Copper. 0·91 7·05	—	—	—	—	0·37	—	0·39	0·42
Magnesium. Iron. Copper. 0·58 2·09 3·9	0·35	0·36	0·36	0·37	0·37	0·38	0·40	0·41
Iron-Copper-Aluminium.								
Iron. Copper. 0·82 7·91	—	—	—	—	0·35	0·38	0·40	0·42
2·09 8·18	—	—	—	—	0·34	0·36	0·38	0·39
Nickel-Copper-Aluminium.								
Nickel. Copper. 1·12 7·93	—	—	—	—	0·38	0·40	0·41	0·42
2·16 9·04	—	—	—	—	0·36	0·39	0·40	0·41
1·96 11·87	—	—	—	—	0·35	—	0·38	0·39
Manganese-Copper-Aluminium.								
Manganese. Copper. 0·98 13·9	0·26	0·29	—	—	0·27	0·30	0·37	0·38
0·98 8·0	0·25	0·29	—	—	0·25	0·30	0·36	0·39
Manganese-Nickel-Copper-Aluminium.								
Magnesium. Nickel. Copper. 1* 2* 9·08	—	—	—	—	0·33	0·37	0·38	0·39
1·5* 2* 8*	0·34	0·36	0·39	0·40	—	—	—	
1·5* 2* 4*	0·35	0·37	. .	—	0·37	0·40	0·40	0·41
Pure Aluminium	0·50	0·50	—	. .	0·52	0·52	. .	—
Cast iron†	0·102	0·099	0·121	0·113	—	—	—	—

* Nominal composition.

	† Graphite.	Combined Carbon.	Silicon.	Sulphur.	Phosphorus.	Manganese.
As cast	2·64	0·85	1·84	0·078	1·09	0·82
Annealed	3·54	0·15	1·94	0·070	1·09	0·85

although some of the alloys have an appreciably higher conductivity
This applies particularly to alloy ")

II(c). *Heat-Treatment of Castings.*

The very great improvement in the mechanical properties of
a number of aluminium alloys, particularly those containing
magnesium, which can be produced by heating and quenching
from suitable temperatures in the case of forged and wrought
material, obviously suggests the possibility that similar improvement
might be obtainable in castings by proper heat-treatment. The
fact that it has hitherto been thought impossible to obtain such an
improvement in the case of castings made of the composition of
Duralumin, has tended to retard development in that direction
The present Authors, however, having had occasion to apply heat
treatment to cast material of various compositions, soon arrived
at the conclusion that improvement in mechanical properties could
be obtained if the heat treatment were suitably arranged. This
very promising line of work has been pursued to a considerable
extent, and the results described below have been obtained.

In the case of the simple binary alloys of copper and aluminium,
it might not at first sight be anticipated that heat-treatment could
produce any very marked results. In connexion with an inquiry
into the so-called " burning " of aluminium copper alloy pistons,
however, it was decided to ascertain if possible the mechanical
properties of a saturated solid solution of copper in aluminium.
Work on the constitution of these alloys has previously shown that
the solubility of copper in solid aluminium is of the order of 5 per
cent, but that this amount of copper, or rather of the copper-
aluminium compound $CuAl_2$, could only be brought into solid
solution by prolonged annealing. Normally, in the production of
copper-aluminium chill castings, free compound ($CuAl_2$) is found
in the alloy when the copper content is as low as 2 per cent. For
the purpose of the present experiment, chill-cast bars containing
2 per cent, 3 per cent, 4 per cent and $4\frac{1}{2}$ per cent of copper were
prepared. These were annealed for three days at a temperature
between $450°$ and $520°$ C. After this treatment, microscopic

examination showed that no undissolved compound $CuAl_2$ was present, except in the case of the alloy containing $4\frac{1}{2}$ per cent of copper, where a minute quantity could still be found. Results of tensile tests on these alloys are shown by the thick lines in the graphs

FIG. 22.

Tensile Tests on Heat-Treated Chill Castings of Copper-Aluminium Alloys.
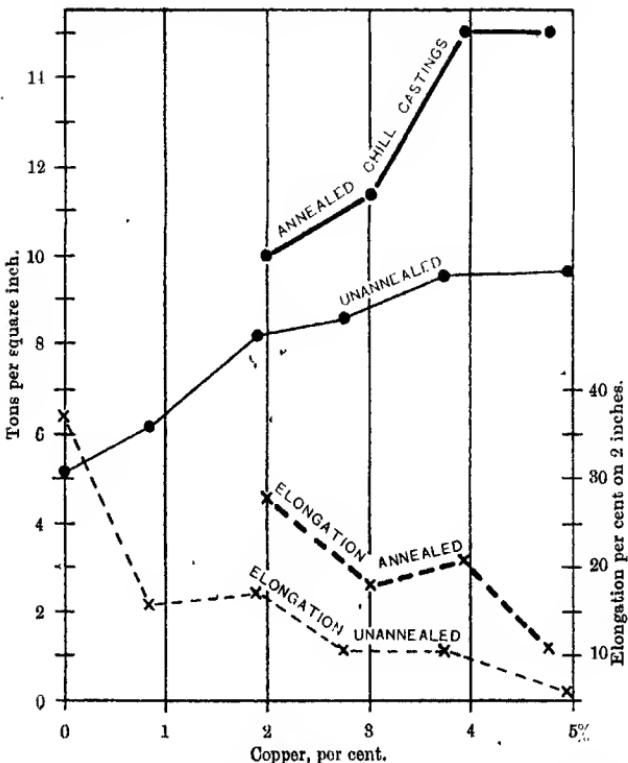
of Fig. 22. For comparison, the results of similar tensile tests on unannealed chilled castings of the same composition are shown by the thin lines in the same Fig., the data being taken from the Eighth Report to the Alloys Research Committee. The result of heat-treatment in this case has been to raise the ultimate stress to a very marked

degree ; the improvement in the alloy containing 3 per cent copper is from 8 tons per square inch to 10 tons per square inch, and in the alloy containing 4½ per cent of copper, from 9½ tons to 15 tons per square inch. This very remarkable increase in tensile strength is accompanied by an appreciable increase in elongation. The results in regard to elongation are also shown in the graphs of Fig. 22. These results indicate very clearly that considerable improvement in the mechanical properties of copper-aluminium alloys may be obtained from suitable heat-treatment, applied to castings.   It appears probable that still higher tensile strength could be obtained by the use of alloys containing a higher percentage of copper, which would leave an appreciable amount of the hard stiffening body, $CuAl_2$ in the alloys even after the prolonged annealing described above.

Still more striking and important results have been obtained by heat-treatment applied to castings of the alloy "Y" (4/2/1½). The heat-treatment applied to this alloy is similar in certain respects to that which has been found successful when applied to the rolled material, but in order to bring the alloy in the first instance into a condition resembling as nearly as possible that of the wrought material, a prolonged heating at a temperature in the neighbourhood of 500° C. is applied.   In the earlier stages of this work, the temperatures employed were restricted to the neighbourhood of 500° C., in view of the fact that in the case of alloys similar to Duralumin a temperature appreciably higher than 510° C. is known to damage the material. Subsequent work on this alloy, as has been indicated below, has shown, however, that much higher temperatures can safely be employed, and both heating and quenching at and from temperatures as high as 530° C. have been successfully used.

The results obtained are indicated in Table 4. It will be seen that the heat treatment there described consists in heating for six hours at 480° C. either in an open muffle or in a salt bath (nitrate), followed by quenching in water. Quenching in boiling water or in oil has also been used and gives on the whole somewhat better results than quenching in cold water. The data given in the

Table are the ultimate stress results obtained in the material both
as cast, and after heat-treatment, when tested both at the ordinary
temperature and at 250° C. In addition, in order to ascertain the

TABLE 4.

Alloy " Y." 1-inch diameter Chills.

Treatment.	Ultimate Stress. Tons per square inch.	
	15° C.	250° C.
As Cast	12·6	10·5
Tempered 48 hrs. at 250° C. . . .	9·8	8·6
HEAT TREATED.		
(1) Open Muffle:—		
6 hrs. 480° C. quenched and aged . .	14·9	11·3
,, ,, tempered 48 hrs. at 250° C.	13·1	10·1
48 hrs. 480° C. quenched and aged . .	15 8	12·0
,, ,, tempered 48 hrs. at 250° C.	13·8	9·4†
(2) Nitrate Bath:—		
6 hrs. 480° C. quenched and aged . .	14·6	12·8
,, ,, tempered 48 hrs. at 250° C.	13·2	11·0
48 hrs. 480° C. quenched and aged . .	16·0⁴	13·7
,, ,, tempered 48 hrs. at 250° C.	16·0	12·0
(3) Nitrate Bath:—		
6 hrs. 530° C. quenched in boiling water and aged . . . . . ⟩	21 0‡	—

* Maximum ext. at 15° C. 2·25 per cent on 2 inches.

† Maximum ext. at 250° C. 3·75 per cent on 2 inches.

‡ Extension at 15° C., 6·5%.

effect of prolonged heating at 250° C. on alloys which have been
heat-treated in the manner described, tests have also been made
both at the ordinary temperature and at 250° C. on specimens

which, subsequent to heat-treatment, have been tempered by
heating for 48 hours at 250° C.

It will be seen on inspection of the Table, that heat-treatment
produces very important improvements, in the tensile strength of
the material, both when tested at the ordinary temperature and
at 250° C. The effect of prolonged heating at 250° C. upon the
tensile strength both at the ordinary temperature and at 250° C. is
very marked on the material as cast. After heat-treatment, the
effect of prolonged heating at 250° C. appears to be very
considerably less. In a heat-treated casting of alloy " Y " it will be
seen that the tensile strength of 16 tons per square inch is
unaffected by two days' tempering at 250° C., although it falls
to 12 tons per square inch if tested at 250° C. The latter figure,
however, must be compared with 10·5 tons per square inch,
which represents the tensile strength of the material as cast and
tested at 250° C. The elongations, as indicated in the footnote to
the Table, are practically unaffected by heat-treatment.   The
effect of tempering and of testing at 250° C. material which has
been heat treated at 530° C. has not yet been fully tested.   It is
interesting to add, that heat-treatment of this kind is now carried
out in regular industrial practice at the Royal Airship Works at
Cardington, and that an ultimate tensile stress as high as 20 tons
per square inch was first obtained there. The importance of so
marked an improvement in the strength of castings of alloy " Y " by
a comparatively simple heat-treatment is, of course, obvious.   It
should open up a field of application for aluminium alloy castings
which the relatively low strength of such material has hitherto
rendered impossible.

In connexion with the heat-treatment of castings of alloy " Y,"
the microstructures of the material in the various conditions have
been carefully studied. These are illustrated in Figs. 23 to 28, Plate 3,
Fig. 23 shows the alloy as cast, unetched, under a magnification
of 150 diameters. The same structure after etching in hydrofluoric
acid, which darkens the aluminium-nickel compound $(NiAl_3)$
is shown in Fig. 24 under a magnification of 400 diameters.
The same structure etched with hot nitric acid, which probably

darkens the copper compound ($CuAl_2$) is shown in Fig. 25 also under a magnification of 400 diameters. Fig. 26 shows the structure of the cast material after heating for six hours at 530°. C., quenched in cold water. This section had not been etched, and is shown under a magnification of 150 diameters. It will be seen that the amount of free compound present is materially less than that shown in Fig. 23 prior to heat treatment. Fig. 27 shows the same structure as Fig. 26 under a higher magnification, 500 diameters. The very large effect which is produced on the structure of the annealed and quenched alloy by re-heating it to 500° C. followed by slow cooling is illustrated in Fig. 28 under a magnification of 500 diameters. Here a large amount of a dark etching constituent is present. The alloy in this condition has, of course, been completely softened and will undergo no appreciable hardening by subsequent ageing. This micrograph clearly illustrates the fact that the age-hardening property of the alloy after heating and quenching is due to the retention, by quenching, in solid solution of one or more of the compounds present. As indicated in the section on the constitution of these alloys, the subsequent age-hardening is no doubt due to the separation of this dissolved compound in a very fiuely-divided state

Section III.

Wrought Alloys.

The present Section deals with alloys capable of being wrought by such operations as forging, rolling, extrusion, cold drawing, etc., and relates not only to the mechanical properties of the alloys when heat-treated, but also to their behaviour while undergoing working and describes the methods which have been studied and adopted for overcoming the numerous and serious difficulties which were met at every stage in endeavouring to forge and roll alloys which had hitherto been regarded as unsuited for treatment of that kind.

III(a). *The Ternary Alloys of Aluminium, Zinc and Copper.*

In view of the results and anticipations indicated in the Tenth Report to the Alloys Research Committee, the first step undertaken was a preliminary exploration of the behaviour and properties of the ternary alloys of copper and zinc with aluminium. Remarkable results already obtained with one of these alloys (3/25) were given in an Appendix to the Tenth Report, but before proceeding to a more detailed study either of that alloy or of any other member of the group, it became desirable to ascertain the range of properties which might become available by variations of composition in either direction This preliminary work, carried out before the experimental rolling-mill at the National Physical Laboratory was

Fig. 29.
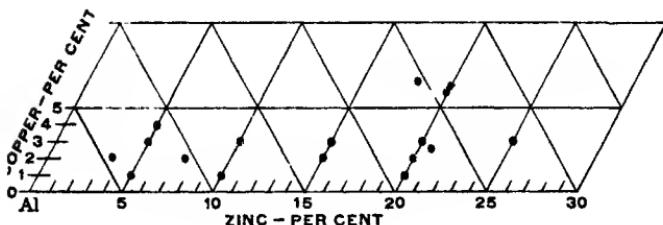
available, will only be briefly described here, mainly for the purpose of showing, first, the very great difficulties which were encountered so long as rolling experiments had to be carried out in an industrial plant under conditions which could not be easily regulated to suit the materials under trial, and in the second place, in order to record interesting data which have been obtained in regard to a number of alloys which may prove to be of some permanent interest, although they have not been chosen for further careful study in the present research.

The ternary alloys of copper and zinc with aluminium which were chosen for this preliminary work, are tabulated in Table 5 which gives both the composition aimed at and the actual composition of the alloys as found by analysis. The ternary alloys tested are

hown in their relation to the basal diagram of the ternary system
n Fig. 29. In addition to these, a number of special alloys

TABLE 5.

Composition of Ternary Alloy Billets Rolled.

Melting No.	Composition per cent.					
	Intended.		Found by Analysis.			
	Zinc.	Copper.	Zinc.	Copper.	Iron.	Silicon.
H 1	5	4	4·64	3·92	0·10	0·14
2	5	3	4·71	2·67		—
3	10		9·96	2·75		. .
4	15		14 74	2·50	—	—
5	25		23·48	2·67	—	—
6	5	2	3·5	2·31	..	—
7	10		7·38	1·76		—
8	15		13·99	1·64		
9	20		18·95	1·66	—	—
10	5	1	4 76	0·87	—	—
11	10		9 91	0·94	—	—
12	20		18·45	0·80	0·16	0·12
13	30	3	28·73	2·87		—
4 A	20	3	20 52	3·13	—	—
5 A	25	3	—	3·13	—	—
5 B			25·25	2·90	—	—
5 C			24·47	2·78	.	—

were also prepared in order to test the effect of the addition of
magnesium and of certain impurities. The compositions of these

also are indicated in the Table, but they cannot be represented on the diagram.

The preparation of these alloys was carried out on much the same lines as those adopted in connexion with the work described in the Tenth Report, both the zinc and aluminium used being of the same quality as that described there. The copper employed was remelted refined electro-copper. The materials used for this portion of the research were kindly supplied free of charge by Messrs. The British Aluminium Co., Ltd., as regards aluminium, the zinc by Messrs. Brunner Mond and Co., Ltd., and the copper by the Broughton Copper Co.

The alloys were melted in quantities of from 20 to 25 lb., sufficient to allow the following castings to be made : —

(1) One 3 inch diameter chill cast billet, 19 inches long.

(2) Three chill castings, 1 inch diameter by 7 inches long, and

(3) Four shaped sand castings.

The melting was done in salamander crucibles, closed with a lid, in an oil-fired injector furnace. No charcoal or flux of any kind was used. The aluminium was melted first, the copper added next as 50 per cent copper-aluminium alloy, and the zinc added last. The melts were stirred with a graphite rod. In weighing out the alloys no allowance was made for loss of any of the various metals ; in some of the earlier melts the compositions found by analysis showed somewhat considerable variations from those aimed at. With later melts, however, where the temperature conditions were regulated by means of pyrometers in the molten metal, more satisfactory results were obtained. In these preliminary melts, pouring temperatures were not determined, but merely judged by eye.

The rolling of the alloys was carried out at the rolling mills of the British Aluminium Co., at Milton, Staff., by the kind permission of the Company. Every possible facility and assistance was given to the Authors in carrying out this work, in spite of the fact that it necessitated some interruption of the regular production work of the mill. It was not, however, possible to alter the speed of rolling

or the passes provided in the rolls, and these were all adjusted to
suit the very soft material normally treated, namely, pure aluminium.
These rolling experiments were carried out at various dates between
July, 1910, and April, 1913. Before the latter date, the Authors had
reached the conviction that success in work of this kind could only
be assured if rolling operations could be carried out in a special
experimental mill under their own control, and the erection and
installation of the necessary plant at the National Physical
Laboratory had at that time been begun.

Rolling and Drawing Operations at Milton.

The procedure adopted was the same as that described in the
Tenth Report, namely, preliminary breaking down, hot, from
3 inches diameter to $1\frac{1}{4}$ inch diameter. The rods so obtained
were then divided, two 18-inch lengths being cut from each and
reserved for drawing operations, while the remainder was reheated
and further hot-rolled to rods of $\frac{7}{8}$ inch diameter.

In the case of the first series of billets rolled (H1 to H12,
Table 5), no attempt was made to determine the actual temperature
of the billets during hot-rolling. The billets were preheated in
the manner usually employed in the case of pure aluminium, to a
temperature of about 400° C. in one of the ordinary muffles used for
aluminium.

With the exception of the hardest alloy, H5 (3/25), no serious
difficulty was experienced in breaking down to $1\frac{1}{4}$ inch diameter
rods. In the case of the 3/25 alloy, a portion of the billet broke up in
the early stages, its behaviour suggesting that too high a breaking
down temperature had been employed. On allowing the remaining
sound portion to cool somewhat, it was successfully rolled to $1\frac{1}{4}$ inch
diameter.

On a later occasion, when melts H13, 4A, 5A, 6B and 5C, were
rolled, the temperatures of the billets were measured by thermo-
couples. In order to do this, without boring a hole in the billet,
a small aluminium casting having a saddle shape and grooved on its
under side, was placed on the billet and heated with it as shown

in section in Fig. 30. Before rolling, the billet was drawn to the front of the muffle and its temperature determined by inserting the thermo-couple in the groove between saddle and billet. The use of this device led to a distinct improvement in the breaking down of the harder alloys 3/25 and 3/20. It was soon found that in the case of these alloys, temperatures considerably below 400° C. were most favourable for breaking down, and temperatures determined in the above manner, as low as 300° C. were satisfactorily employed.

Fig. 30.

Scale.—Half size.
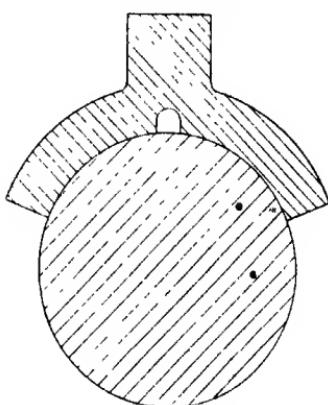
The behaviour of other billets which gave interesting results is indicated in the following tabulation :—

H5 A (3/25)   Rolled at 395° C.—Billet fell to powder at the first pass.

H5 C (3/25)   Rolled at 300° C.—Rolled to 1¾ inch—fin cut off, rod re-heated and rolled to 1-inch diameter. 18 inches re-heated and rolled to ⅞-inch.

H13 (3/30)   Rolled at 290° C.—Rolled to 1¼ inch, but showed bad cracking

H4 A (3/20)   Rolled at 310° C.—Rolled to 1¼-inch diameter. 18 inches re-heated and rolled to ⅝-inch diameter.

In addition to the billets which were intended for rolling into rods, a certain number of slabs had also been cast in a sand mould,

measuring $13\frac{1}{4}$ inches by $10\frac{1}{4}$ inches, with a thickness of $1\frac{1}{8}$ inch, the materials used being H5 (3/25) and H4 (3/15). These were broken down at 300° C. and 290° C. respectively to 9 gauge (0·144 inch). The blanks thus obtained were divided into two, and one-half of each annealed at 400° C. and subsequently cold rolled to 15 gauge (0·072 inch). The finished sheet showed considerable surface spill and serious edge cracking. Tests of this sheet in the case of the 3/25 alloy have already been given in the Appendix to the Tenth Report, page 143.

Cold Drawing.

Attempts at cold drawing were made on the hot-rolled rod obtained from the experiments described above, both without any annealing, and with preliminary and intermediate annealing at about 400° C. In the absence of annealing very limited success was obtained. Only the alloys of low zinc and copper content could be reduced to $1\frac{3}{8}$ inch diameter. The alloys 4/5, 3/5, 3/10, 3/15, 2/20 and 1/20 broke before this stage could be reached.

When annealing at 400° C. was used, all the alloys from H1 to H12, with the exception of 3/25, were drawn in two draughts to 1 inch diameter, after a preliminary annealing. After a further annealing, drawing was continued and was successfully accomplished to $1\frac{3}{8}$ inch diameter except in the cases of alloys 4/5, 3/15, 2/20 and 1/20, which drew hollow and broke. Further cold-rolling from $1\frac{1}{8}$ inch diameter rod was attempted on a later occasion with the alloy 3/20 (melt H4 A) and 3/25 (melt H5 B). These and subsequent efforts, however, failed until at a very much later stage, when, as the result of work carried out with the experimental mill at the National Physical Laboratory, the properties of these materials were better understood. Ultimately, the 3/20 alloy "A" was successfully drawn from 0·5 inch diameter rolled rod to 13 gauge wire by Messrs. Henry Wiggin and Co., Ltd., at their Birmingham works.

Density.

The density of the ternary copper-zinc-aluminium alloys studied in the form of $\frac{7}{8}$-inch hot-rolled rod has been determined at atmospheric temperature by the Metrology Department of the

Laboratory. The results for the 1 2 and 3 per cent copper series, and for the alloy containing copper 4, zinc 5 per cent respectively, are shown in the graphs of Fig. 31. In the case of other

FIG. 31. *Density.*

Grammes per cubic centimetre.

⅜-inch Hot-Rolled Rod.
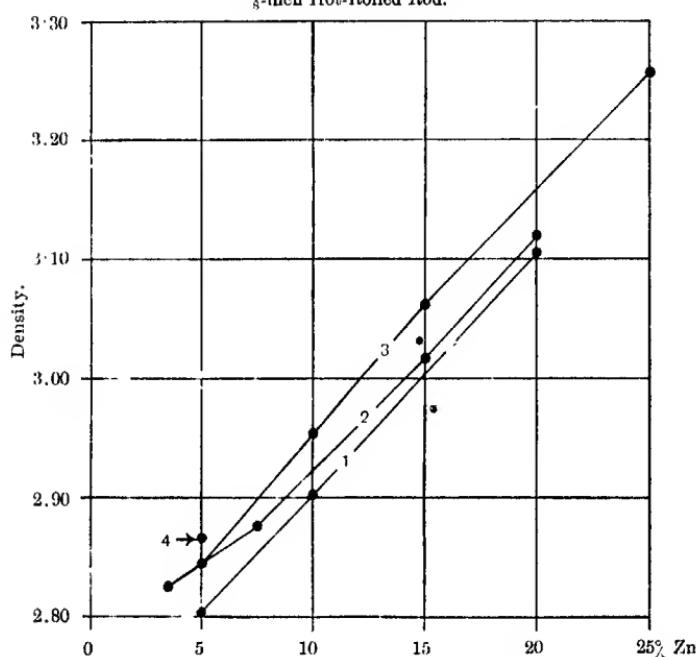
wrought alloys studied, results are included in the summary Tables of Section III (e).

Mechanical Properties.

The complete study of the mechanical properties of the material obtained from the rolling experiments described above, has never been carried out, owing to conditions which have already been referred to. The tests which have actually been carried out are indicated in

TABLE 6.—*Summary of Tests made.*

Page Reference	Melt No.	1	2	3	4	5	6	7	8	9	10	11	12	13	4A	5A	5B	5C
	Nominal Composition { Copper	4	3	3	3	3	2	2	2	2	1	1	1	3	3	3	3	3
	{ Zinc	5	5	10	15	25	5	10	15	20	5	10	20	30	20	25	8	25
	Tensile Tests—																	
53	1¼ in. rod—hot-rolled	×	×	×	×	—	×	×	×	×	×	×	×	×	×	—	×	×
54	⅞ in. rod—hot-rolled	×	×	×	×	×	×	×	×	×	×	×	×	—	×	—	×	—
56	½ in. rod—hot-rolled	—	—	—	—	—	—	—	—	—	—	—	—	—	×	—	×	—
56	1¾ in. rod—hard-drawn	×	×	×	×	—	×	×	×	×	×	×	×	—	—	—	—	—
56	Material drawn from ½ in. rod	—	—	—	—	—	—	—	—	—	—	—	—	—	×	—	×	×
	Elastic Properties—																	
58	⅞ in. rod—hot-rolled	—	—	×	×	—	—	×	—	×	—	×	×	—	—	—	—	—
58	1¾ in. rod—hard-drawn	—	—	×	—	—	—	×	—	—	—	×	×	—	—	—	—	—
	Hardness—																	
62	Brinell	×	×	×	×	—	×	×	×	×	×	×	×	—	—	—	—	—
62	Scleroscope	—	—	×	×	—	—	—	—	×	—	×	×	—	—	—	—	—
60	Arnold Test	—	—	×	×	×	—	—	—	—	×	—	—	—	—	—	—	—

Table 6, where page references to results of the various tests are included For the purpose of completing .ó a *certain extent the available data, a few of the tests referred to in Table 7, and quoted in detail below have been carried out in the present year (1920), nearly 10 years after the material was rolled. This fact is of some importance as showing the properties of material which

FIG. 32.—*Tensile Tests on Hot-Rolled Rods.*
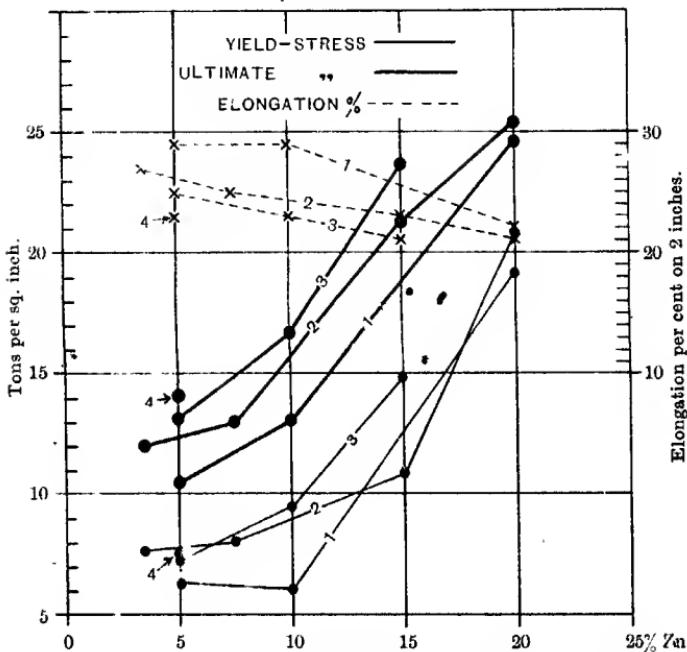
1¼-inch diameter.

has undergone the ageing process, to which reference is made below, to a very complete extent.

Tensile Tests on Hot-Rolled Rods 1¼ inch Diameter
and ⅞ inch Diameter.

The results of tensile tests carried out on test-pieces of standard shape (B.E.S.A. Test-Piece C) with screwed ends, are shown in the

graphs of Figs. 32 and 33. In these graphs the zinc content of the alloys is used as abscissa and the tensile strength as ordinate, while the points plotted from the results of series of alloys having the same copper content are united by the lines forming the graphs. There are, therefore, graphs representing alloys with

FIG. 33.—*Tensile Tests on Hot-Rolled Rods.*

⅛-inch diameter.
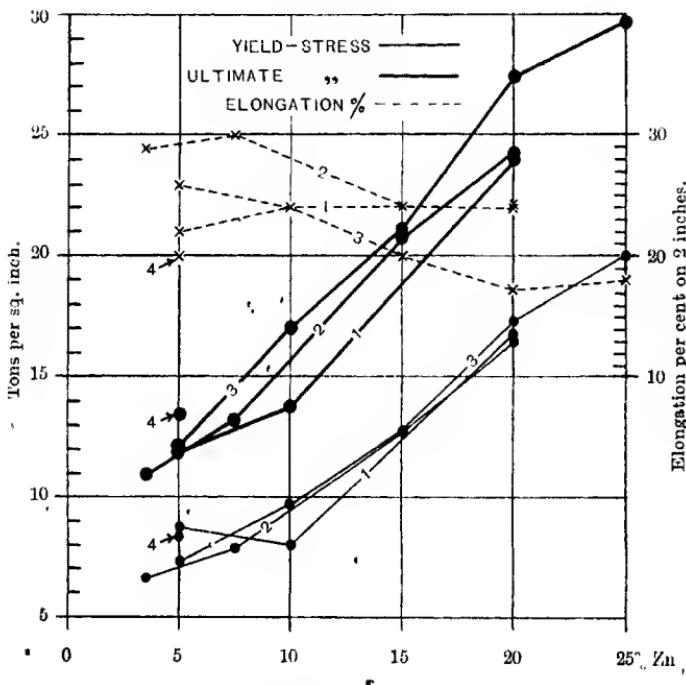
varying zinc content and containing respectively 1, 2 and 3 per cent of copper. In regard to alloys containing 4 per cent of copper, one only was successfully rolled, so that there is a single point for each property of this alloy instead of a graph. In order to diminish the number of separate graphs on the same Figure, the results for reduction of area are plotted in separate Figures, Nos. 34 and

35. For purpose of comparison, however, the results from pure aluminium-zinc alloys are shown on the latter diagrams.

It will be seen that the graphs for the two series of rods are very similar, but there is some superiority in elongation in the case of the $\frac{7}{8}$-inch diameter material. Tests on rod $\frac{1}{2}$ inch diameter are

Fig. 34.— *Reduction of Area. Hot-Rolled Rod.*

1¼-inch diameter.
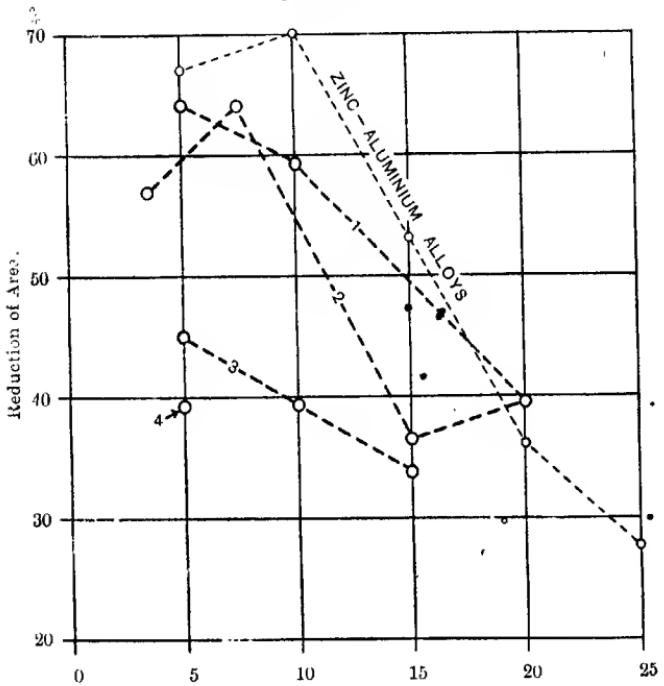
only available in the case of alloys 3/20 (H4 A) and 3/25 (H5 B). These show a slight superiority in ultimate stress and a considerable increase in elongation over the $\frac{7}{8}$-inch rod. The results are shown in Table 7.

All the graphs indicate, as might be expected, an increase of hardness with increase of copper content. Copper is found to be

a more potent hardening agent than zinc, and this fact is well brought out in the graphs of Figs. 36 (page 60) and 37 (page 62), where the specific tenacities for the 3 per cent copper ternary

TABLE 7.

Tensile Tests—Hot-Rolled to ½-inch Diameter.

Alloy.			Yield stress.	Ultimate stress.	Elongation per cent on 1 inch.
			Tons per square inch.	Tons per square inch.	
3/20 (H4A)	.	. .	21·2	27·6	25
3/25 (H5B)	.	. .	27·1	31·7	21

alloys in the form of ⅝-inch diameter hot-rolled bars are plotted against density and zinc content respectively.

TABLE 8.

Tensile Tests on Material Hard-Drawn from ½-inch Diameter.

Material.	Yield Stress.	Ultimate Stress.	Elongation per cent on 2 inches.
	Tons per sq. inch.	Tons per sq. inch.	
3/20 Alloy.			
(H4A) Drawn ½-³⁄₁₆ inch. .	32·8	34·1	7·5
3/25 Alloy.			
(H5B) Drawn ½-⁷⁄₁₆ inch. .	24·4	32·8	11
(H5C) Drawn ½ ⅜ inch. .	31·0	33·3	12·5

Tensile Tests on Hard-Drawn Material.

Tensile tests on the hard-drawn material have been carried out in a similar manner to that employed in the case of the hot-rolled rod, but, as already indicated, a number of these rods broke or drew

hollow during the process of production, so that the results must be regarded as of doubtful value.

Some data representing the results of tensile tests on material obtained from the attempt to draw $\frac{1}{2}$-inch rod down to lower diameters are shown in Table 8.

Fig. 35.—*Reduction of Area. Hot-Rolled Rods.*

$\frac{7}{8}$-inch diameter.
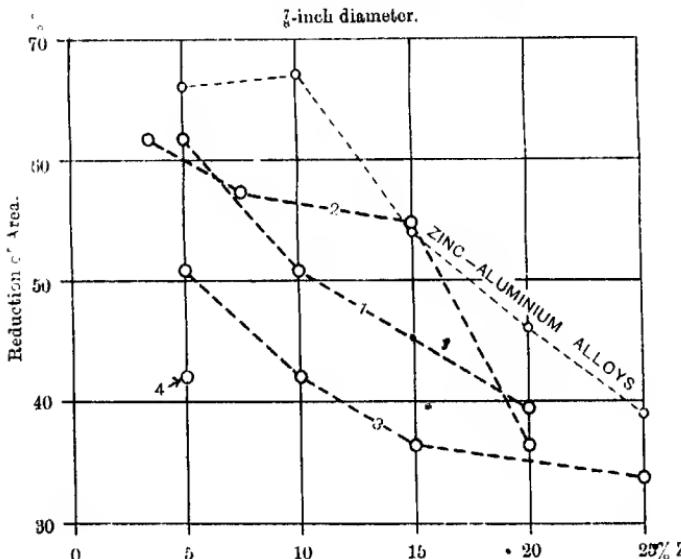
Elastic Properties of the Alloys.

Determinations of the elastic modulus and the elastic limit (limit of proportionality) on the material now under consideration have only been carried out in the present year (1920), nearly ten years after the alloys had been rolled. These determinations were carried out in the Engineering Department of the National Physical Laboratory, and were made on test-pieces 0·5 inch diameter and 8 inches between gauge marks, the material used being both hot-rolled $\frac{7}{8}$-inch diameter and hard-drawn $1\frac{3}{8}$ inch diameter rod. Stress-strain diagrams and elastic curves derived from these observations se

far as they relate to the hot-rolled material are shown in Figs.
38 (page 63) and 39 (page 64). The values found for elastic limit
and modulus are given in Table 9 where the alloys are arranged
in the order of increase in elastic limit.

The values of the yield and ultimate stress, elongation and
reduction of area obtained from these 8-inch test-pieces are given

TABLE 9.

Elastic Properties— $\frac{7}{8}$-inch diameter Hot-Rolled and $1\frac{1}{16}$-inch diameter Hard-Drawn Rod.

Alloy.		Elastic Limit.	Elastic Modulus.
Hot-Rolled.		Tons per sq. inch.	Lb. per sq. inch.
2/10	(H7R)	5·5	$10·00 \times 10^6$
1/10	(H11R)	8·0	9·67
3/10	(H3R)	8·7	9·74
3/15	(H4R)	12·0	9·74
2/20	(H9R)	12·2	9·67
1/20	(H12R)	13·1	9·57
Hard-Drawn.			
2/10	(H7D)	8·7	9·88
3/10	(H3D)	9·3	9·45
1/10	(H11D)	10·0	9·74
1/20	(H12D)	11·8	9·51

in Table 10 in the columns headed " Ten years," these words
indicating the age of the specimens on which the tests were carried
out. For purposes of comparison, in an adjoining column are stated
the corresponding data from tests on 2 inch test-pieces made at a
time when the same alloys were only seven weeks old.

The above Table shows the important fact that the tensile
properties of this series of alloys in the hot-rolled condition have

undergone a material improvement in the course of ten years' ageing. This improvement takes the form of a marked increase in ultimate stress and yield without a serious reduction of ductility. The elastic limit values obtained are high in comparison with accepted

TABLE 10.

½-inch Diameter Rod. Aged 7 Weeks and 10 Years.
Hot-Rolled and Hard Drawn.

Alloy.	Yield. Tons per sq. inch.		Ultimate. Tons per sq. inch.		Elongation per cent on 2 inches.		Reduction of Area per cent.	
		10 years.		10 years.		10 years.		10 years.
Hot-rolled –								
2/10 (H7) .	7·8	9·5	13·1	14·9	30	27	57	56
1/10 (H11) .	7·4	11·7	13·7	17·0	26	27	51	54
3/10 (H3) .	9·7	13·5	17·1	19·6	24	24	42	42
3/15 (H4) .	13·6	–	21·6	24·2	22	22	36	39
2/20 (H9) .	16·4	–	24·2	26·0	24	20	36	29
1/20 (H12) .	16·5	23·5	24·2	26·2	24	20	39	29
Hard drawn—								
2/10 (H7) .	14·4	15·0	16·9	17·9	13	10	—	26
3/10 (H3) .	7·8		14·6	12·3	5	1	–	15
1/10 (H11) .	13·3	15·8	15·2	18·5	15	14	—	33
1/20 (H12) .	20·1	24·5	23·8	25·1	9	4	—	15

values of other wrought aluminium alloys of similar density in this condition as hot-rolled.

Further Tests on the Alloys.

It had originally been intended to carry out the systematic testing of these alloys in the very complete manner which has been applied to the alloys of aluminium and zinc as described in the

Tenth Report. In view of the evident superiority of certain of
the alloys, this very extensive program has never been completed.
Certain additional tests were, however, carried out, and the results
are recorded below.

Fig. 36.— *Specific Tenacity.*
⅞-inch diameter Rods.
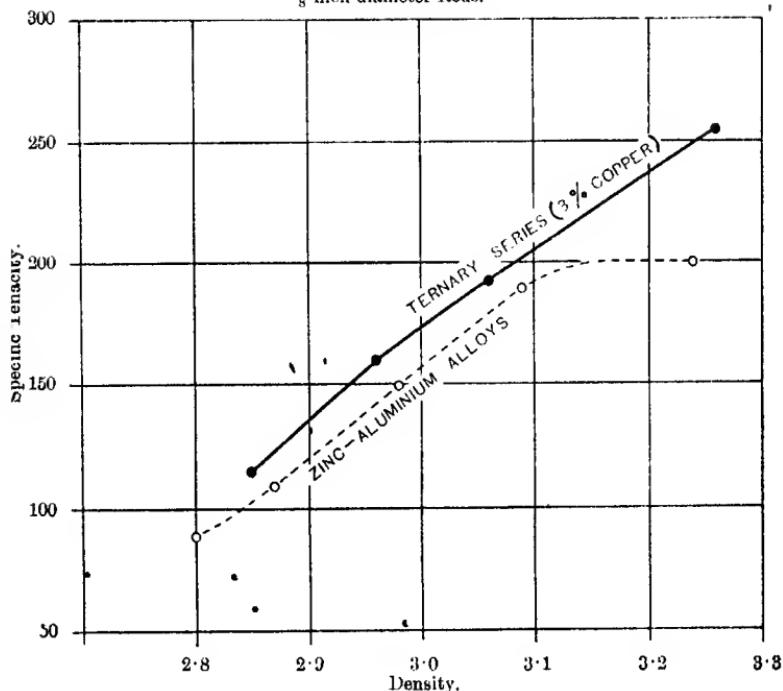
Alternate Bending Tests (Arnold).

A series of test-pieces of these alloys, similar to those prepared
from the aluminium-zinc series described in the Tenth Report
were tested by the kindness of Professor J. O. Arnold, F.R.S., in
his special alternate-bending testing-machine at Sheffield University.
The tests were carried out at the same time as those of the aluminium-
zinc alloys, and Professor Arnold's remarks relating to these tests,

TABLE 11.

Arnold Tests.

Alloy.	Alternations Endured.	Mean.
	54	
	56	
3 20 (H4)	52	52·5
	48	
	40	
	38	
3/25 (H5)	40	40·0
	42	
	60	
	62	
1/5 (H10)	60	61·0
	62	
	58	
	68	
3/10 (H3)	62	63·0
	64	
	52	
	62	
3/15 (H4)	52	56·0
	58	

as quoted on p. 434 of the Tenth Report, apply equally to the present series. The results are contained in Table 11 (page 61).

Hardness Numbers.

The hardness numbers of the alloys from H1 to H12, see Table 5 (page 46), have been determined both by the Brinell and Scleroscope

FIG. 37.—*Specific Tenacity.*

$\frac{3}{4}$-inch diameter Rod.   Hot-Rolled.
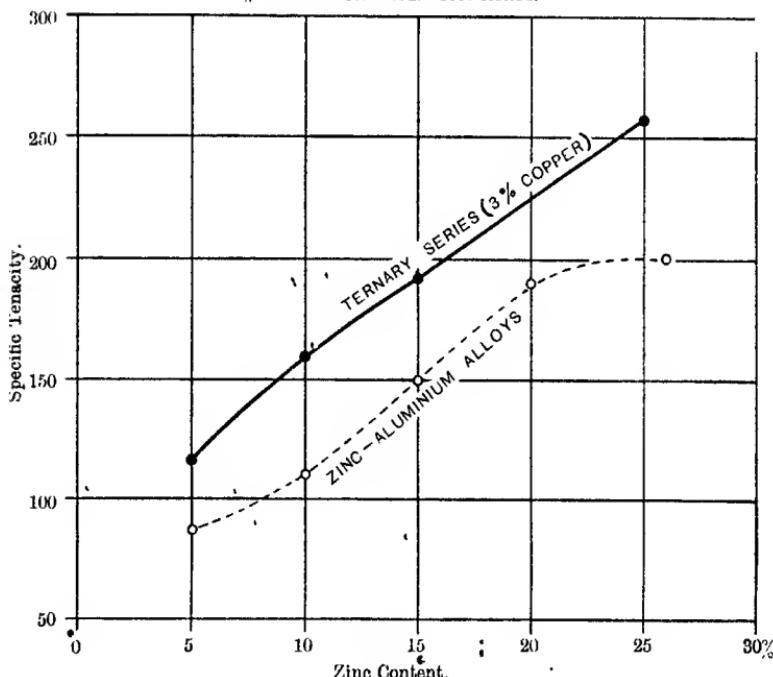
methods, the tests being made on transverse sections of hot-rolled rods 1$\frac{1}{4}$ inch diameter. The Brinell tests were made with a 9·52 mm. ball and a load of 1,000kg., while the Scleroscope readings were obtained with the universal hammer. The results are shown in the graphs of Fig. 40 (page 65). The results shown in that Figure afford

FIG. 38.—*Elastic Limit.* " H " *Series.   Rolled Bars.*

ALLOY
H12
H9
HA
H3
H11
H7

[Stress in tons per sq. inch.]

Extension.

20 18 16 14 12 10 8 6 4 2 0
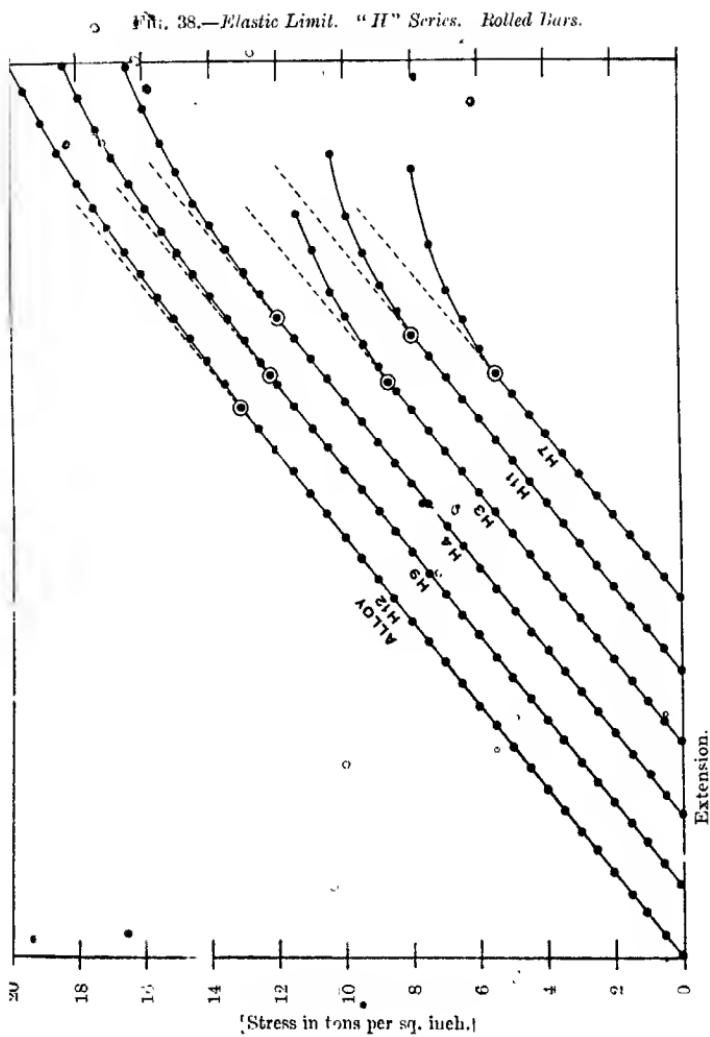
Fig. 39.
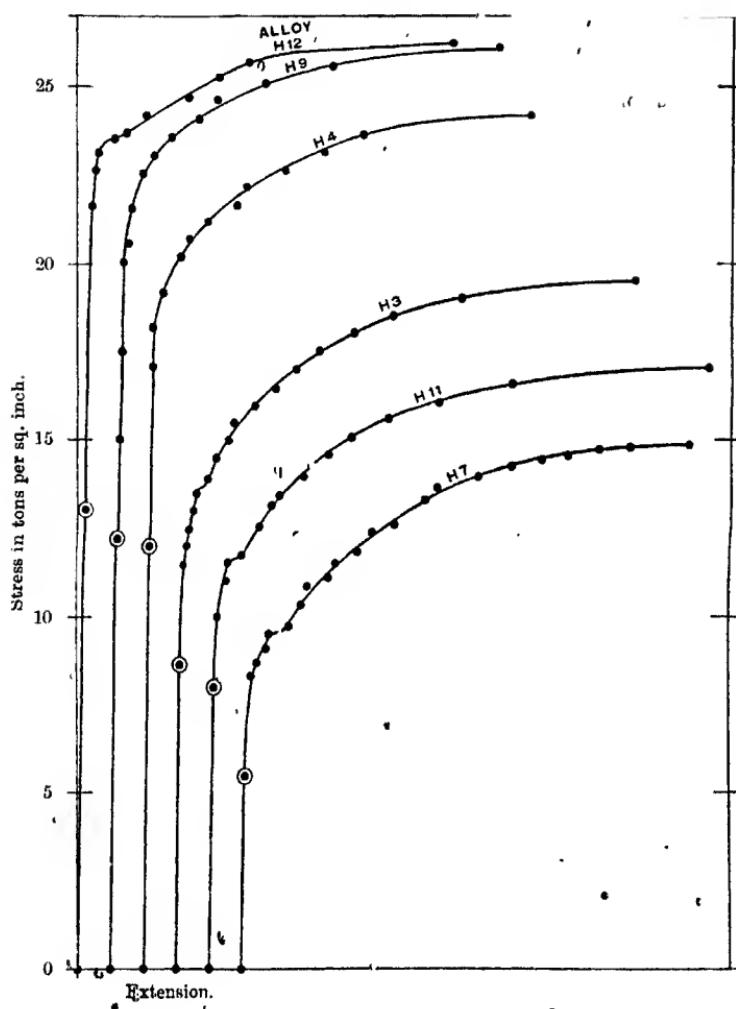
interesting comparison with the results of a similar series of tests made on samples of the same material after ten years' ageing,

FIG. 40.--*Hardness*. •
1¼-inch Hot-Rolled Rod.
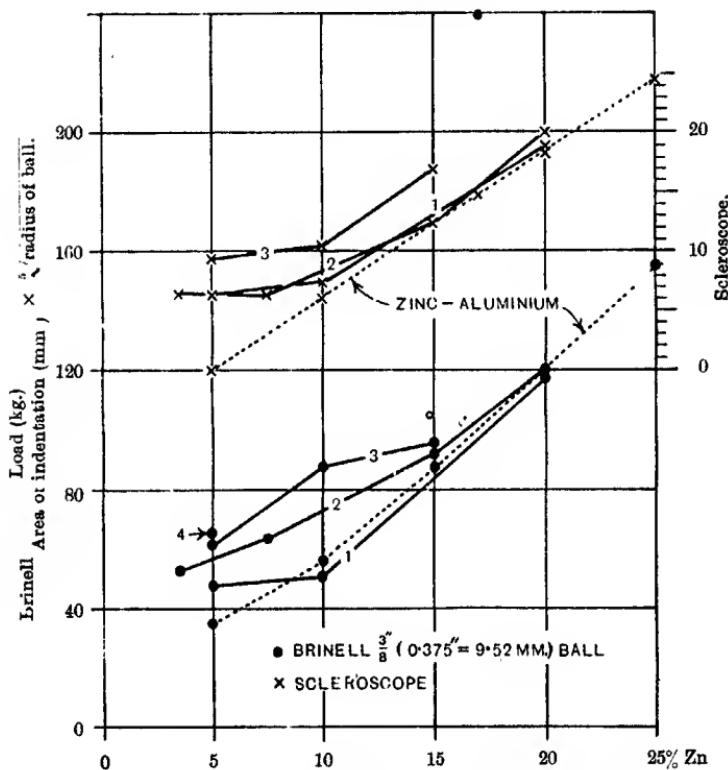
which are given in the section dealing particularly with that subject Section IV(*a*), (p. 155).

Microstructure of the Alloys.

The microstructures of these alloys present no very striking features, other than those already met with in the alloys of zinc and aluminium, with the exception that the presence of the copper-

aluminium compound, $CuAl_2$, makes itself felt by the presence of a dark etching constituent. The study of the equilibria of the ternary alloys, as described in Section V., p. 196 of the present report, indicates that the copper compound in quantities corresponding to 3 per cent of copper should, in equilibrium, disappear from the microstructure, passing into solid solution in the aluminium. As actually manufactured, however, the alloys always show the presence of the dark-etching copper compound. The structures actually met with are illustrated in the four micrographs of Plate 4. Figs. 41 and 42 relate to alloys containing 5 per cent of zinc and 1 per cent and 3 per cent of copper respectively. The increase in the amount of the dark etching constituent in Fig. 42 as compared with 41 is evident. Similarly Figs. 43 and 44 relate to alloys containing 20 per cent of zinc and 1 per cent and 3 per cent of copper respectively. All the micrographs are shown under a magnification of 150 diameters.

Further Study of Selected Alloys.

The material obtained as the result of experimental rolling described above, and the results of tests made upon that material, furnished the Authors with a series of data which clearly indicated the existence of an important and promising group of alloys of the ternary system, copper-zinc-aluminium. In many cases the material in question was not entirely above suspicion as regards its quality from the rolling point of view, and only occasional billets of some of the alloys had been successfully rolled. The results as they stood, however, were sufficient to indicate that the harder alloys of the group, such as 3/25 and 3/20, possessed very remarkable properties, combining with a reasonable degree of ductility a specific tenacity quite as high as, or even slightly higher than, that of the best aluminium alloys available at that time.[*] These copper-zinc-aluminium alloys possess the further advantage that they require no heat treatment, such as quenching and ageing, in order to develop their best qualities. It was also hoped from the outset that one or other of these alloys might form the basis

for further advance by the addition of other elements. Experimental billets containing various added elements in quantities of 1 per cent and notably a set of billets containing $\frac{1}{2}$ per cent of magnesium had been prepared in the course of the preliminary work, and an attempt had been made to roll them at Milton. With the exception of a billet containing 1 per cent of iron, however, all these experimental materials broke up completely in the rolls. The Authors, none the less, felt confident that this early failure resulted from causes which could be overcome by a careful systematic study of the alloys and of the methods to be employed in rolling them. In spite of the fact that they were assured by some of those most conversant with the properties and behaviour of aluminium alloys, that materials containing as much as 20 per cent of zinc must be regarded as suitable only for casting, and that such alloys could never be commercially rolled, the Authors considered that more exhaustive study would enable them to overcome these difficulties. They may, perhaps, be permitted to claim that their confidence in this respect has been fully justified by the results actually achieved. The fundamental fact remained at the outset that on several occasions billets of these alloys *had* been successfully rolled, so that, provided the conditions which had accidentally occurred on those particular occasions could be studied and determined, systematic success should be obtainable.

Thanks to the installation at the National Physical Laboratory of a well-equipped foundry and rolling-mill specially designed for experimental work on light alloys, the Authors have been enabled to undertake the systematic study of the rolling of the harder alloys of zinc, copper, and aluminium, as well as of other alloys which are referred to in a later Section of this Report. As a definite result achieved by this study, they are now able to report the development of a series of aluminium alloys, some of which possess properties which are as yet quite unrivalled by any similar products developed elsewhere Not only have these alloys been developed and produced on an experimental scale in the Laboratory, but their production has been successfully transferred to a full-scale plant operating under industrial conditions, and successful production in that plant has

been rapidly and easily achieved on the basis of the results already obtained at Teddington. In addition to the development of a definite series of alloys, a large amount of valuable information has been obtained in regard to the general problems of rolling some of the more difficult alloys of aluminium, and incidentally also in regard to their behaviour on casting.

The systematic work in question, although rendered possible by the special and almost unique equipment which was placed at the Authors' disposal for the purpose, was carried out under war conditions which in many respects very much retarded its progress, the systematic work being frequently interrupted to allow of the solution of other problems which appeared to be more immediately pressing. The amount of work actually carried out was, however, very large, and within the scope of the present Report it is not possible to do more than to record it in very brief outline.

The Experimental Foundry and Rolling-Mill.

The experimental foundry where the melting and casting of all experimental alloys has been carried out, occupies a floor space of 60 feet by 27 feet. A general view of the interior is shown in Fig. 45, Plate 5. The metal-melting furnaces, which include one natural draught furnace of the Richmond type, a blast-driven injector furnace by Fletcher Russell, and a mould-heating stove, occupy a trench which is covered over by a steel grating in such a manner that the top openings of the furnaces are flush with the floor. The remainder of the foundry is occupied by recuperative gas-fired furnaces of a larger size, and by electrically heated furnaces. The experimental meltings of aluminium alloys have usually been carried out in the Richmond natural draught furnace. A pyrometric outfit is provided, whereby the temperature of the metal, both during melting and prior to casting, can be determined. There are also electrical connexions to the accurate temperature-measuring apparatus situated elsewhere in the department.

The moulds employed for casting material intended to be used for forging and rolling, are either circular cast-iron chills, 3 inches diameter and 19 inches long, or slab moulds of various sizes ranging

from 12 inches by 5 inches by ¾ inch thick to 14 inches by 7 inches by 1¼ inch thick. In connexion with the casting of the larger slabs, where it was found desirable to tilt the mould during pouring, a special mould-tilting machine has been designed, the actual construction having been carried out at the Cardington Airship Works. This appliance, with a slab mould in place, is shown in the photograph Fig. 46, Plate 5.

A general view of the experimental rolling-mill is shown in Fig. 47. Plate 6. It consists essentially of a pair of housings capable of accommodating a two-high stand of rolls measuring 30 inches in width and 15 inches in diameter. Originally two sets of bar-rolls for breaking down and finishing, respectively, having a " Gothic " section were provided. Subsequently a set of bar-rolls having approximately circular grooves were also provided, as well as a set of breaking down and finishing rolls for sheet rolling. By means of the travelling crane and some special appliances, roll changing is considerably facilitated and can be carried out in approximately two hours. The rolls are driven through the usual tumbler joints and gearing by means of rope drive from a 60 H.P., D.C. electric motor, which is, however, capable of withstanding 100 per cent overload for a short time. The rope drive from the motor goes through a 7½-ton fly-wheel 8 feet in diameter. The electrical power for driving this mill is derived from a large storage battery. This form of drive was adopted because the local power supply company declined an experimental rolling-mill as a load on their ordinary lighting circuit, it has, however, the very great advantage that by varying the voltage applied to the motor, the speed of the mill can be regulated in any desired manner, and switchboards for this purpose are provided. In this manner the linear velocity of the rolls can be varied from 50 feet to 200 feet per minute.

One point which required special consideration in an experimental mill of this kind was the temperature of the rolls themselves. In regular industrial working, the rolls attain a steady and fairly high temperature, when hot-rolling is being carried out, simply by the supply of heat which they receive from the slabs or billets which are passed through. In an experimental mill, where only a few

pieces are rolled at any one time, it becomes necessary to heat the
rolls to a suitable temperature before rolling is commenced. The
arrangement finally adopted for this purpose consists of two rows
of gas-burners, which can be temporarily placed so that the flames
play on the rolls while these are slowly revolved. The temperature
attained by the rolls is measured by means of a thermometer inserted
into a small hole drilled into the shoulder of the roll about 0·5 inch
below the surface. The arrangement is shown in place with a pair
of grooved rolls in Fig. 48, Plate 6.

In addition to the rolling-mill proper, the building contains
several other appliances, including an electrically-heated pre-heating
furnace. A general view of this furnace is shown in Fig. 49, Plate 7.
It consists essentially of a steel tube, 12 inches in diameter by 7 feet
in length. This is heated by a resistor formed of nichrome ribbon
wound upon the steel tube, a layer of micanite being interposed.
The whole is insulated with magnesia bricks and contained in an
iron-framed box, sheeted with uralite. The windings are so arranged
that various sections can be connected either in series or parallel
as may be desired. In order to minimize the cooling effect of the
ends of such a furnace, the stoppers which are used for closing the
ends are also provided with electrical heating coils, Fig. 50, Plate 7,
and in this way a remarkable degree of uniformity of temperature
within the furnace is achieved. This furnace was constructed in
the Laboratory in 1915, and has served for the whole work of the
rolling-mill in regard to light alloys for over five years. A total of
well over 1,000 heats have been carried out in it, and the furnace
has so far required no repair or attention at any time. It runs at a
temperature of 400° C. with a current consumption of two kilowatts,
but a larger current consumption is required while the furnace is
being heated up. A much smaller furnace of somewhat similar type,
but capable of heating only one billet, is also provided.

Particular attention was given to the exact determination of
the temperature of billets and slabs when heated for rolling purposes
in the electric pre-heating furnace. The ideal method in every
case would be to insert a thermo-couple in a small hole actually
drilled in the billets or slabs being heated. This, however, is not

possible owing to the damage to the slab or billet which would be caused by such a hole. The plan adopted was therefore to charge into the furnace, together with the slabs or billets under experiment, an exactly similar slab or billet specially prepared and kept for that purpose, which was provided with a suitable hole for the insertion of a thermo-couple. Since the temperature of the whole furnace was kept very uniform, never varying by more than 10° C. from one end to another, it was sufficiently exact to assume that the dummy billet or slab would rise in temperature at approximately the same rate as the others charged with it, all being cold when first introduced into the furnace. This method has proved very satisfactory and has allowed of the exact control of the temperatures of billets when used for forging or rolling.

The additional appliances available include in the first place a small forging press capable of exerting a maximum pressure of 80 tons. Although forging by the press has not proved very satisfactory, this appliance has none the less proved particularly useful in the straightening of bars and other objects which are liable to become bent during rolling.

For purposes of forging, two power-hammers have been installed. The first one installed is a belt-driven hammer of the oscillating-spring type. This has proved extremely useful and the great bulk of the work described in the present Report has been forged with this implement. It has, however, proved to be somewhat insufficient in weight and power for the larger slabs and billets which have had to be dealt with. More recently, a second and somewhat more powerful pneumatic forging-hammer has been installed. This was purchased by the Laboratory, at a specially reduced price, from the late Professor A. K. Huntington at a time when the latter was dispersing the equipment of the former metallurgical department of King's College, University of London.

Production of Light Alloy Bars and Rods.

After preliminary work on some of the softer and more easily rolled alloys, a systematic endeavour was made to roll billets of

the alloy "B" (3/25) into bars and rods. Although the Authors had become aware that preliminary forging appeared to be necessary for the successful rolling of the harder aluminium alloys the power hammer was not yet available when this work was commenced, and initial attempts were made at forging under the press. The results obtained however, were very variable and unsatisfactory, although some fifty billets 3 inches in diameter of the 3/25 alloy were employed. A number of other devices were also tried, including variation of the working temperature from 400° C. down to 325° C., as well as variations in the speed of rolling, at the breaking down stage, from 45 feet to 180 feet per minute. Variations in the severity of pinch were also tried but did not succeed in eliminating the serious cracking which always occurred. Prolonged annealing at a temperature in the neighbourhood of the solidus of the alloys was also carried out on a batch of billets previous to rolling, but again no marked success was obtained, and only a few rods, 1-3/8 inch diameter, were successfully rolled.

In view of the circumstances that the production at an early date of a satisfactory light alloy was at that time regarded as of urgent importance, the Authors considered that although the alloy "B" (3/25) should ultimately be found capable of regular satisfactory production, the problem appeared to be too lengthy and difficult to be pursued at that time. Comparison of the available data showed that the distinctly softer alloy containing copper 3 per cent and zinc 20 per cent (alloy "A") would fulfil most requirements at least as well as the harder and slightly heavier B (3/25), and attention was thereafter concentrated upon the development of the alloy "A" (3/20).

III (b).   *Development of Alloy "A"* (3/20).

Between 28th June 1915 and 10th March 1916, attention was devoted to the production of rolled rods from 3-inch billets of the 3/20 alloy, and more than one hundred billets were cast and rolled. Satisfactory breaking down without surface cracking was achieved and successfully demonstrated to representatives of the industry interested in the production of light alloys. In the earlier stages

of this work, the attempt was made to roll billets of this alloy
without preliminary forging, but the results were not encouraging.
As soon, therefore, as the power-hammer was available, experiments
on forging under the hammer were carried out. It was found that
at temperatures between 350° C. and 400°,C. the billets would
stand a moderate amount of forging without cracking. At first
this forging was attempted by using blows distributed over the
whole of the surface while the billets were held in heavy tongs and
rotated about their axis. Forging in this manner was found to
produce cracks and splitting at the ends, and in subsequent work the
billets were not rotated during hammering. Blows were distributed
from end to end, the billet being moved parallel to its axis. The
billet was then turned through 90°, and an equal number of blows

FIG. 51.
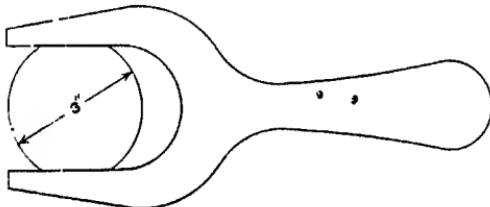
applied in that direction. Satisfactory forging was obtained
by this means, and the amount of forging was then carefully
regulated by the use of saddle-shaped gauges, of the shape indicated
in the sketch Fig. 51. In the method finally adopted, the
amount of reduction applied by forging lay between 5 and 10 per
cent, and the billets were forged to an octagonal section in two
stages with intermediate reheating.

Although the forging operation was successful in itself, that is,
it did not lead to immediate crackin gof the billets, most of the
billets thus treated cracked seriously during breaking down in the
rolls. The results obtained were inconsistent, satisfactory and
unsatisfactory rods being obtained indiscriminately. Attention
was then given to a number of factors, which, it was thought, might
affect the result. These included questions of the quality of the

metal used in making the alloys, the severity of the pinch, speed
of rolling, and the use of higher temperatures, such as 400° C. in
forging. Variation of these factors produced no improvement, nor did
microscopic examination of the billets as cast, and after forging and
rolling, throw any light on the difficulty. The view was, therefore,
thrust upon the Authors that the source of the trouble probably
lay further back, namely, in the casting operation. Very careful
attention was, therefore, given to the method of casting the billets
in the foundry.

Careful examination of the billets as cast, and at an early stage
of the rolling when cracking had commenced, led the Authors to
suspect that surface defects in the castings, giving rise to weakness
of the skin, led to the development of cracks during breaking down.
The effect of forging appeared to consist to a large extent in closing
up and obliterating these surface defects, but if these were at all
serious the amount of forging which was applied was obviously not
sufficient to avoid consequent cracking. Careful study was, therefore,
devoted to the question of producing billets free from surface defects.
This was ultimately achieved by adopting a method of pouring the
liquid metal into the mould as slowly as was possible without leading
to the formation of cold shuts and other defects. For this purpose
the billet moulds were tilted to an angle of about 45° to the horizontal
and the molten alloy was poured in as a slow stream. The object
of this procedure was to maintain in the mould at any one time as
shallow a layer of liquid metal as possible, the rise in the liquid
metal being arranged to take place at about the same rate as the
solidification of the alloy. Pouring in this manner results in the
practical abolition of all shrinkage cavities and largely avoids the
necessity of " following up." The adoption of these methods,
coupled with the use of moulds at a relatively low temperature
(100° C.), brought about an immediate change for the better in the
behaviour of the billets in the rolling-mill.

Subsequently further modifications in treatment were investigated.
These included first, annealing the billets after forging and before
breaking down in the mill at as high a temperature as possible
short of incipient fusion (about 510° to 520° C.). In some cases

this treatment was applied between the first and second forging operations. The object was to remove as completely as possible the coring of the solid solution crystals in the cast material, and to bring about the absorption of the free compounds, particularly $CuAl_2$, into the solid solution. Microscopic examination of the billets so treated showed that the greater part of the compound $CuAl_2$ had passed into solution, and that vigorous grain growth had taken place. Photo-micrographs illustrating this effect are shown in Figs. 52 to 54, Plate 8. Fig. 52 shows the material as cast ; Fig. 53 shows it after forging and annealing at 400° C. Fig. 54 shows the structure of the billet after forging and annealing at 400° C., and further drastically annealed at 500° to 520° C. All these photomicrographs are under a magnification of 150 diameters. It is interesting to note that hardness tests (Brinell) show no increase of hardness consequent upon the absorption of the compound. It was hoped, by obtaining in the forged and annealed billets a microstructure very similar to that of the rolled bars, that a considerable increase in ductility of the material would be obtained. Unfortunately, this treatment did not produce the desired effect.

The use of a lower breaking down temperature (300° C.) was then tried, but no improvement was found.

Finally, completely successful results were obtained by increasing the amount of forging and employing a higher forging temperature (400° C.). This modification of the treatment was coupled with intermediate annealing between the first and second forging operations and final annealing for one hour at 450° C. before rolling. In addition, it was found necessary to keep the rolls thoroughly well lubricated. This latter precaution was essential, since it was found that dry clean rolls allowed the hot aluminium alloy to seize in contact with the roll surface, and when this occurred the billets were split in half and curled around the rolls instead of being formed into bars.

It was found that when, by the methods just indicated, the billets had been broken down and rolled into rods $1\frac{5}{8}$ inch or $1\frac{1}{4}$ inch diameter, there was no difficulty in rolling these rods further to $\frac{7}{8}$ inch diameter or smaller. For this purpose it was only necessary to re-heat the rods as produced from the breaking down process

to a temperature in the neighbourhood of 400° C. : at that
temperature they could be rolled to any desired extent, high speed
and heavy pinches being employed without fear of producing cracks

Physical Properties of Alloy " A " (Rods).

The results of tensile tests on Hot-Rolled Rods of Alloy " A '
(3/20) in various sizes are given in Table 12.

TABLE 12.

Diameter of Rod (hot-rolled).	Yield Stress. Tons por square inch.	Ultimate Stress. Tons per square inch.	Elongation por cent on 2 inches.
1½ inch.	17·8	25·4	16·5
	17·7	26·9	18·5
1¼ inch.	18·5	27·4	18·0
	16·2	26·1	19·0
⅞ inch	16·5	26·1	22·0

The results given in this Table are in all cases mean values of at
least two duplicate tests and really represent typical values taken
from long series of tests. It will be seen that the properties of the
alloy do not vary very much with the diameter to which it has been
reduced by hot-rolling, once the ingot, cast to 3 inches in diameter,
has been reduced to a diameter of 1½ inch. In addition to
the results of ordinary tensile tests as shown in Table 12,
determinations of the Elastic Limit (Limit of Proportionality) and
of the Elastic Modulus (Young's Modulus) have also been made ;
the Limit of Proportionality, in rod ⅞ inch in diameter was found, in
two different batches of the alloy, to lie at 11·8 and 13·0 tons per
square inch. This value may be contrasted with the figure given for
the " yield " in tests on the same material above, namely, 17·7 and
17·8 tons per square inch respectively. The elastic modulus for
this material was found to be $9 \cdot 7 \times 10^6$ lbs. per square inch. The

results of other series of tests on Alloy "A" in rod form are given below in other connexions.

Production of Sheet.

The problem of breaking down and rolling billets of the alloy " A " (3/20) into rods having thus been successfully solved, the question of the production of the alloy in the form of sheets was next attacked. In view of the fact that for many of the most important uses of light alloys, these materials are required in the form of sheets or of strips suitable for forming into angles and channels of various shapes, this problem assumed very considerable and urgent importance.

The first systematic work on the production of sheets was done on a series of chill cast slabs measuring 5 inches by 12 inches by ¾ inch thick. In accordance with the experience already obtained in the treatment of billets intended for the production of rods, these slabs were subjected to a preliminary forging before rolling. This was done at 400° C and comparatively light blows were used, giving a reduction in thickness of 10 per cent, which was the largest amount permissible. For rolling, the forged slabs were pre-heated to 350° C. and later to 400° C., while the rolls themselves were heated by means of gas-burners to a temperature between 75° and 100° C. This pre-heating of the rolls was intended to minimize the chilling effect of the large mass of cold iron on the small slabs. In industrial practice, the rolls would become heated to a moderate temperature by constant contact with a succession of slabs.

In the earlier stages of this work the slabs were merely rolled in one direction, but later it was found necessary in order to secure greater ductility in the transverse direction to employ cross rolling. In spite of the prevalent opinion that cross rolling of aluminium and its alloys is not feasible, this procedure gave excellent results.

The principal difficulty met with lay, much as in the case of rod rolling, in the occurrence of surface and edge cracking. In addition, low ductility, particularly in a transverse direction, was also encountered.

The surface cracking was again directly traceable to casting defects in the slabs. Observation showed that wherever a slab

showed defects, such as those arising from " drawing," cracking immediately resulted. By adaptation of the methods already devised for the successful casting of billets, slabs having smooth bright surfaces were produced, with the result that surface cracking and edge cracking were practically eliminated.

The low ductility found in the rolled sheets as first produced was

TABLE 13.

Melt No.	Thickness inches.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation per cent. on 2 inches.
Early material showing low ductility.				
W220	0·067	16·1	28·7	9
219	0·073	23·4	29·4	9 5
220	0·035	25·8	30·3	7
Material from slabs rolled without crossing, showing low transverse ductility.				
W250	0·125 *L	19·1	26·3	15
—	†T	19·0	25·3	8
Later material from forged slabs employing cross rolling.				
W303	0·051 *L	16·4	28·0	18
—	†T	21·3	29·0	16
310	0·056 *L	17·3	28·8	20
—	†T	16·7	26·9	15

* Longitudinal. † Transverse.

traced to cold work which was put upon the metal during the rolling owing to the fact that the whole of the operation was conducted at too slow a speed, allowing the metal to cool down too far. By increasing the speed and by limiting the amount of reduction applied between successive reheating, satisfactory ductility was obtained. A series of typical test results illustrating the various stages of progress which have been briefly outlined above are given in

Table 13. The development there indicated represents the progress made as the result of casting and working about 30 slabs of the size named. ·

Once a slab had been satisfactorily broken down from the cast and forged condition into a blank of about one-half the original thickness, no further difficulty was experienced in rolling such blanks down to sheet or strips of any desired gauge. The small size of slab employed, however, did not permit of the production of sheet in sufficient length for practical purposes, and as it was feared that the use of larger and thicker slabs might introduce difficulties of an entirely different nature, it was thought necessary to carry the work further and to study the casting, forging and rolling of larger slabs.

The size of slab adopted measured 7 inches by 14 inches by 1½ inch

Fig. 55.—*Slab Mould with V-Shaped Groove.*

Half-size
Section of
Groove.
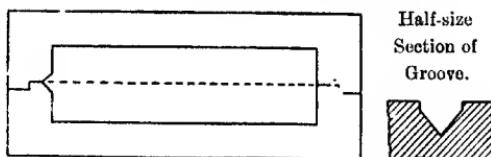
thick. The satisfactory breaking down of these thicker slabs proved a much more difficult problem. In the early stages cracking of the surface and edges was encountered to a far greater extent, and work on some fifty or sixty of these thicker slabs was required before these difficulties could be overcome. The main difficulty resided in obtaining thoroughly satisfactory castings. The fact that the condition of the surface layers of the cast slabs was largely responsible for cracking which occurred during breaking down was shown by results obtained in forging and rolling two slabs whose cast surfaces had been removed in a planing-machine to a depth of approximately ⅛ inch. After forging, these machined slabs gave practically perfect rolled blanks.

In order to overcome the casting difficulty, various mechanical aids were tried. The most successful of these was the provision of a small V-shaped groove cut from top to bottom at one side of

the slab mould as indicated in the sketch, Fig. 55, which shows a horizontal section through the mould, while the V groove is shown at the side half full size. The small V-shaped runner which is thus left on the edge of the slab must, however, be removed before the slab is forged or rolled. If this is not done, the small portion of projecting metal remains in the weak and brittle condition as cast and is unaffected by the earlier stages of forging and rolling. It is, therefore, unable to accommodate itself to the extension of the slab as a whole and leads to very serious cracking.

When by these means, together with careful attention to casting detail, satisfactory slabs had been obtained, it was found possible to break them down into blanks practically free from defects by very slight modifications of the forging and rolling processes already employed for the smaller slabs. The further reduction of these blanks down to sheet of 18 gauge thickness offered no special difficulties.

Cold Working and Spinning of Sheet of Alloy "A" (3/20).

In spite of the relatively high strength and stiffness of "A" sheet in the hot-rolled condition, it was thought desirable to investigate the possibility of utilizing this material for cold working and spinning. At first very considerable difficulties were met with, but these have been overcome by the use of a special annealing temperature which leaves this alloy in a very soft condition.

Some early attempts at spinning this material were made at the Royal Aircraft Establishment (then the Royal Aircraft Factory) at Farnborough. Only a limited amount of success was attained, but the annealing temperature employed, 400° to 450°, was much too high. Some of the spinnings thus produced, after a short period of storage, spontaneously developed cracks. This behaviour is in accord with what is now known to be typical of this material in regard to season cracking when it has been annealed at too high a temperature (see page 175).

In order to arrive at the best conditions for cold working of alloy "A" (3/20) sheet, a series of cold-rolling experiments were carried out. A strip 3½ inches wide by 4 feet 6 inches long of the hot-rolled

material was employed.   Its initial thickness was 0·117 inch, and this was reduced by three stages of cold-rolling with intermediate annealing to a final thickness of 0·048 inch (18 s.w.g.).   The three stages consisted in reduction from (a) 0·117 inch to 0·089 inch, (b) from 0·089 inch to 0·061 inch, and (c) from 0·064 inch to

Fig. 56.
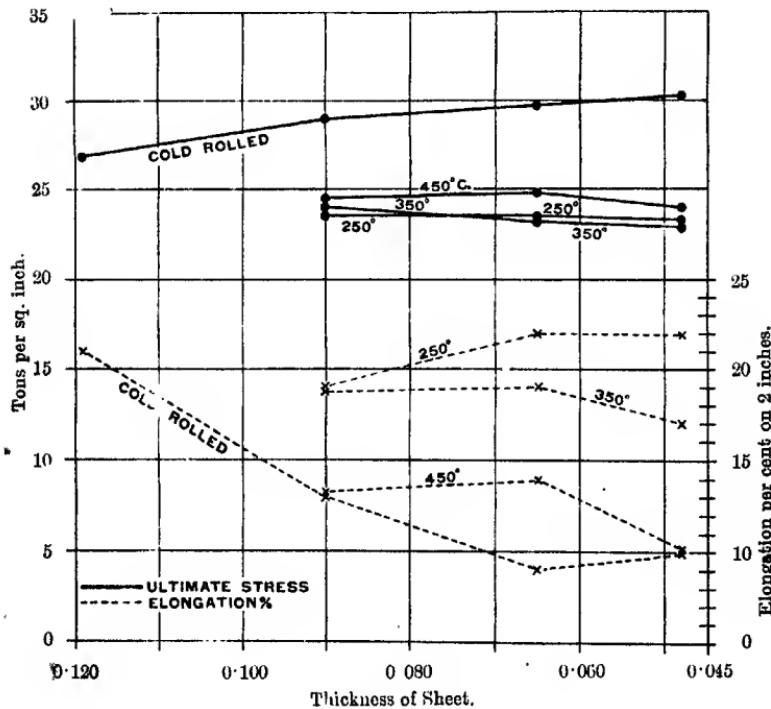
0·048 inch   At each stage a short length was cut from the strip and tensile tests were carried out on the material in the condition as cold-rolled, and after annealing for thirty minutes at temperatures of 250° C., 350° C. and 450° C. respectively, followed by air-cooling. The results of the tensile tests made on these samples as regards ultimate stress and elongation are shown in the graph of Fig. 56.

These results show clearly the superiority of 250° C. as an

annealing temperature as compared with a higher temperature such as 450° C., particularly in restoring the ductility of cold worked strips. It will be seen that annealing at 450° C. leaves the elongation of the thinner strip practically unaffected, while annealing at 250° C. restores the elongation to 19 and 22 per cent respectively.

TABLE 14.

Reduction. (Actual thickness of sheet). Inches.	Yield stress. Tons square inch.		Ultimate stress. Tons square inch.		Elongation per cent on 2 inches.	
	Cold-Rolled.	Annealed 250° C.	Cold-Rolled.	Annealed 250° C.	Cold-Rolled.	Annealed 250° C.
Intermediate annealings at 250° C.						
0·119 to 0·88	29·1	13·4	29·9	23·9	10	19
0·088 to 0·65	24·2	13·1	25·2	—	10	20
0·65 to 0·50	24·9	10·3	25·6	23·8	10	20
0·50 to 0·36	- -	12·7	28·3	—	6	18
0·36 to 0·30	24·1	12·5	26·2	23·8	10	16
0·30 to 0·22	27·5	11·4	28·2	22·9	4	14
Intermediate annealings at 450° C.						
0·122 to 0·93	26·8	10·7	28·2	23·7	9	17
0·93 to 0·65	28·3	10·2	31·2	23·3	9	10
0·65 to 0·49	24·2	11·8	30·8	22·9	7	8
0·49 to 0·37	28·4	26·6	31·9	30·2	4	4
0·37 to 0·28	29·4	14·1	32·2	22·3	4	7

The advantage of using an annealing temperature of 250° C. as compared with 450° C. has been further established by a series of tests made on samples taken from hot-rolled strips which had been subjected to cold-rolling in progressive stages amounting to a 25 per cent reduction in thickness at each stage, with intermediate annealings of thirty minutes at 250° C. and 450° C. respectively. The

results of tensile tests made on samples taken at successive stages are shown in Table 14. It will be seen that with annealing at 450° C. the material hardens up very quickly and ductility is not fully restored between stages of cold-rolling. It is interesting to note, however, that cold-rolled and annealed material does not

FIG. 57.—*Tensile Tests on Annealed Cold-Rolled Sheet.*
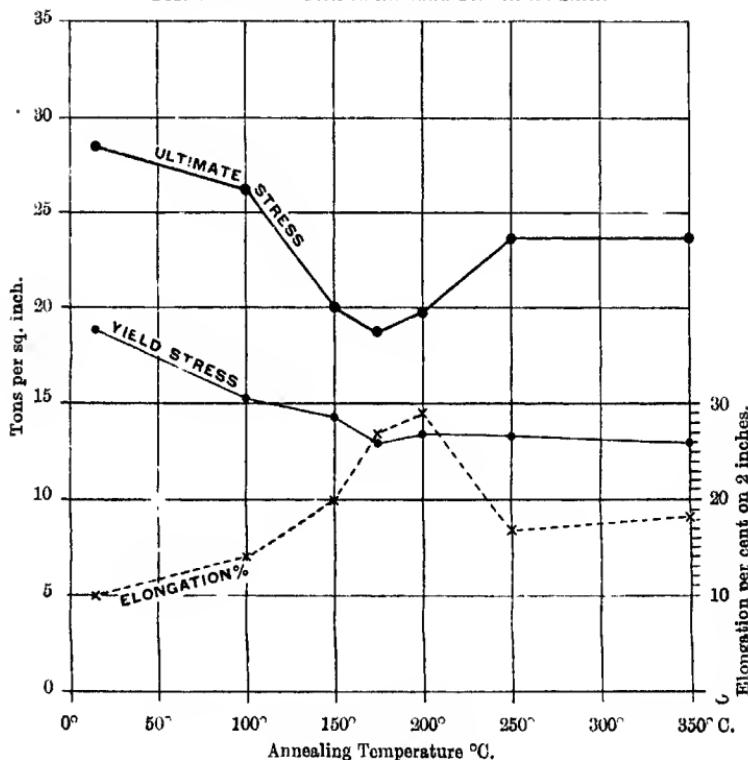
attain to the combination of tensile strength and elongation which can be readily obtained in hot-rolled material which has never been subjected to cold work.

The results above described have been further confirmed by the production of actual cup-shaped spinnings of the alloy "A" (3/20),

G 2

using annealing at 250° C. The spinning itself was readily
accomplished and spinnings made in this way have proved to be
entirely stable, several examples having remained perfectly sound
over a period of four years

Some further tests have been made to determine the effect of
annealing at still lower temperatures than 250° C., and somewhat
remarkable results have been obtained. These are indicated in
the graphs of Fig. 57, which show that a small amount of softening in
cold-rolled sheet becomes apparent after thirty minutes " annealing "
at 100° C., while a temperature of 180° C. brings about the maximum
amount of softening, leaving the material with an extension on
2 inches of as much as 29 per cent and a tensile strength as low as
19 tons per square inch. It is particularly striking to find that
annealing at slightly higher temperatures does not produce so great
an amount of softening.

Production of Very Thin Sheet.

The great possibilities of the alloy " A " (3/20) as regards working,
are well illustrated by the success which has been obtained by rolling
it into very thin sheet. The object aimed at was to test the possibility
of utilizing very thin sheet metal, of low specific gravity, as a
substitute for doped linen fabric for the covering of aeroplane wings
and similar purposes. The weight of such fabric is of the order of
about ½ lb. per square yard, and alloy " A " would have to be reduced
to a thickness of not more than 0·0035 inch to be of equal weight
per square yard. If such material had an ultimate strength of
25 tons per square inch, it would be from three to four times as
strong as the best doped fabric.

The difficulty of producing such very thin sheet in an experimental
rolling-mill, intended for the breaking down and rolling of much
thicker material, was realized at the outset, and the co-operation
of several firms having special rolling plant was therefore secured.
These firms, however, obtained little or no success in their endeavours
to treat this material, although ultimately the British Aluminium
Company at their Warrington Mills succeeded in rolling a strip
of this alloy 7 inches wide, down to a thickness of 0·0025 inch,

producing a material of very beautiful finish and perfectly flat.
For practical use, however, vey much wider strip would be required,
the minimum being about 1½ inches.   A specially designed strip
mill for rolling strips of this width, however, would no doubt lead
to the steady production of the material as required.

Meanwhile, the problem of producing very thin sheet of the alloy
" A " (3/20) was studied by means of the experimental rolling-mill
at the Laboratory.   It was soon found that hot-rolling in these very

FIG. 58.

Scale.—Quarter full size.

HOUSING

CAST IRON

MILD STEEL

BEARING BLOCK
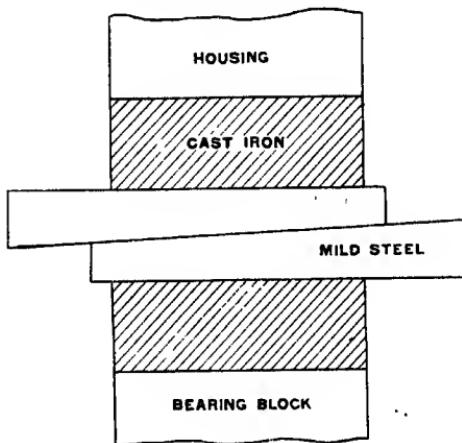
thin sections was impracticable, and cold rolling with annealing
was therefore tried.   No great difficulty was experienced in reducing
the sheet to a thickness of the order of 0·015 inch.   Beyond'
this, however, very great difficulty arose apparently from the
springing or lifting of the rolls.   When attempting to reduce the
thickness below 0·015 inch, the rolls, when idle, ran in close contact.
Even with the adjusting screws hard down, however, there was
sufficient movement to render further reduction very slow.   Part
of this deflection was no doubt due to actual elastic springing of
the rolls themselves, particularly at the neck, but it was evident that

a good deal of displacement took place owing to the insufficiently
rigid manner in which the roll bearings were held in the housings
by means of the screws and blocks provided. These screws and
blocks were, therefore, temporarily replaced by two cast-iron blocks
and two mild steel wedges in the manner shown diagrammatically
in Fig. 58. The use of this rigid arrangement pressing down
on the bearings of the rolls, naturally caused considerable heating.
In these circumstances, therefore, the mill could only be run for a
short time, and had then to be allowed to cool.

This arrangement led to further progress, although at thicknesses
below 0·009 inch it was not possible to obtain more than a reduction
of from 0·001 inch to 0·002 inch in any one stage without reannealing.
The number of passes per stage was even then of the order of twenty.
Further, the greatest care was necessary in keeping the rolls perfectly
clean and well polished, as a patch of oil or grease was liable to cause
destruction of a length of thin sheet in a single pass. Another
difficulty arose from the " camber " of the rolls, which had not
been designed for dealing with such thin material under high pressures.
Fortunately, it was found possible, to some extent, to effect the
necessary adjustment of the roll diameters by means of the row
of gas-burners used for pre-heating the rolls for hot-rolling. By
adjusting the amount of heat supplied to the central or end portions
of the rolls, different amounts of thermal expansion could be brought
about in these portions, with the result that the effect of a greater
or lesser " camber " could be obtained at will.

Ultimately, sheets of the alloy " A " (3/20) were produced of the
following dimensions, 12 inches wide by 6 feet long, thickness
0·005 inch, and 6 inches wide by 15 feet long, thickness 0·0035 inch.
These were flat and free from edge cracks. The guidance obtained
from the production of these very thin sheets in the experimental
mill at the laboratory, subsequently enabled the British Aluminium
Company, at their Warrington mills to produce the longer strip
already referred to.

The results of a series of tensile tests on this very thin sheet are
shown in Table 15, the tests referring both to the material rolled
by the British Aluminium Company and to that produced in the

experimental mill at the Laboratory It will be seen that the alloy preserves a high tensile strength when rolled down to these very thin sections.

In view of the manner in which doped fabrics are damaged by bullet wound, it was necessary to test how perforation by rifle bullets would affect the strength and properties of these very thin alloy sheets. For the purpose of these firing tests, portions of the thin sheet were stretched on wooden frames and rifle bullets were

TABLE 15.

Condition.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation on 2 inches. per cent.	Breaking load per 1-inch width (calculated).
Sheet 0·010 inch thick (British Al. Co.)				
As received	16·5	26·5	8	595
	17·8	26·6	7	
Annealed 250 C	8·7	22 1	11	—
	11·6	22·4	12	
Sheet 0·005 inch thick (N.P.L.)				
Annealed 250° C	--	23·1	2	—
Sheet 0·0035 inch thick (N.P.L.)				
Annealed 250° C	—	23·0	3	180

fired through them both at right angles to the surface of the sheet, and at an angle of 30°. The bullets were found to perforate the sheet without producing any ripping or spreading of cracks from the actual hole. The holes themselves, however, were not circular, but polygonal in shape, consistently eight-sided in the sheet 0·007 inch thick, and seven ded in sheet 0·010 inch thick. The metal appeared to have been cut into a number of converging tongues, which were neatly rolled up behind the corresponding sides of the polygonal hole. The appearance of these holes is illustrated in Fig. 59, Plate 8, showing slightly reduced views of a hole from the entering and

exit sides, respectively. The holes are larger than the bullets, measuring approximately 0·5 inch across from side to side, the bullet used being Mark.VII, S.A.A., 0·303 inch diameter.

The effect of the bullet holes on the strength of test strips 2 inches wide, having the hole in the centre of the width, was determined by means of tests in a machine ordinarily used for testing fabrics. The results are given in Table 16.

TABLE 16.

Tensile tests on wounded sheet, specimens 2 inches wide, 6 inches parallel, bullet hole approximately 0·5 inch diameter.

Rate of loading 200 lb. per minute.

Thickness of sheet.	Breaking load of wounded sheet lb. 2 inches.		Breaking load on 2 inch width calculated from mean breaking load of wounded sheet, assuming a 0·5 inch diameter hole.	
	Sheet as supplied.	Annealed 550° C.	Sheet as supplied.	Annealed 250° C.
0·010 inch	692	592		
	675	595		
	Mean = 684	Mean = 594	912	792
0·007 inch	358	307		
	347	333		
	Mean = 353	Mean = 319	—	—

The breaking loads of the 0·010 sheet per 2-inch width tested in the unwounded state were as follows :—

Sheet as supplied . . . . . 1,000 lb.
Sheet annealed 250° C. . . . . 975 lb.

It will be seen that the strength per unit area of the material after being wounded by a bullet is not seriously diminished. This is in striking contrast to the behaviour of a textile fabric which has been damaged in a similar manner.

Ripping Resistance.

A very serious difficulty which stands in the way of the application of thin sheet for such purposes as aeroplane wing covering, is its relatively low resistance to ripping. Although the tensile strength of a thin metal sheet may be three or four times that of a fabric, the metal sheet can be torn in the fingers much more readily than fabric. This difference in behaviour appears to be due to the circumstance that in a fabric the various fibres are capable of being displaced and deflected in such a manner that a number of them can come into action in resistance to a ripping stress at the same time. In fabrics where the fibres are very firmly held in position, as for instance, in certain dressed silks, ripping is very much easier. As is well known, such materials can be readily torn, whereas softer materials, like woollen fabrics, cannot be readily torn. A continuous sheet of metal may be regarded as the extreme case of a material in which the fibres are very fine and very numerous and are completely attached to one another at every crossing point. Such a material, when exposed to a ripping stress, is torn step by step, and were it not for a certain amount of ductile yielding of the metal, the actual shearing stress at the end of a tear would be infinite even for a small applied load. It is not surprising to find that resistance to ripping depends as much on the ductility as upon the actual strength of the metal. Whether this liability to rip is a serious disadvantage in the case of an aeroplane wing could be tested in actual flight. Under war conditions, jagged rents might easily be made by projectiles. Such rents occur in fabric-covered wings, but the question arises whether the wind pressure on the projecting tongues of metal would be sufficient to cause the tear to spread by ripping. These are, however, questions of aeroplane design and construction to which no further reference need be made here.

Other Properties of Alloy "A" (3/20).

In the course of the Authors' study of the properties of the alloy "A" (3/20), a very wide range of tests has been applied to the material in a variety of conditions. In the majority of cases, the results of

these tests have only been of service in indicating the superiority or otherwise of certain methods of production and treatment. It is, however, desirable to record the results of certain of these tests at least, in order that the fullest information regarding the properties of the material may be available. The tests made include compression, fatigue (alternating stress), and notched-bar impact, at ordinary and at high temperatures.

Compression Tests.

Two sets of tests on rod have been made. In the first series the test-pieces were $\frac{1}{2}$ inch diameter by $2\frac{3}{4}$ inches long, and were held in holders making struts 6·2 inches long. The results are given in Table 17.

TABLE 17.

Material.	Elastic Limit. Tons per square inch.	Buckling Stress. Tons per square inch.	Modulus. Lb. per square inch.
Alloy A (3/20). (1) Extruded.			
(E12) 2¼-inch diameter . .	10·9	13·0	$10·0 \times 10^6$
(E11) 1¼-inch diameter . .	9·0	12·4	$9·8 \times 10^6$
(2) Extruded —Rolled.			
(E12) 1½-inch diameter (from 2½-inch)	10·4	14·3	$9·8 \times 10^6$
¾-inch diameter (from 2½ inch)	11·5	14·6	$9·9 \times 10^6$
(3) Forged and Rolled.			
(W204) 1⅝-inch diameter (from 3-inch billet) . . .	10·6	12·6	$9·9 \times 10^6$
(W209) 1½-inch diameter (from 3-inch billet) . . .	9·5	13·2	$9·7 \times 10^6$
Comparison Alloy (D). Extruded—Rolled.			
(E16) 1½-inch diameter (from 2½-inch)	9·0	13·7	$10·1 \times 10^6$
—	9·2	12·4	$10·3 \times 10^6$

The results given in the last two lines of this Table, under the heading Comparison Alloy "D," are included for purposes of comparison. The alloy in question corresponds in composition and treatment to the well-known proprietary alloy Duralumin.

In the second series of tests, the results of which are given in Table 18, the test-pieces were arranged to have a ratio of length to diameter not greater than 4, so as to give true compression data. Here the comparison material is mild-steel.

TABLE 18.

Material.	Elastic Limit. Tons per square inch.	Buckling Limit. Tons per square inch.	Modulus. Lb. per square inch.
Alloy "A" (3/20). (1) Extruded.			
H E10-2 inches diameter	9·43	—	$9·5 \times 10^6$
Mild-steel	9·4	—	$31·8 \times 10^6$

Some further data on the behaviour of alloy "A" (3/20) under compression are contained in Table 46 (page 152), which relates to a series of compression tests carried out at the Cardington Aircraft Factory (now R.A.W.), on material in the form of channel sections, having a thickness of 0·048 inch and a length of 6 inches. This Table presents a comparison of a whole series of alloys, and the results for the alloy "A" (3/20) are found in the third line of the Table. This Table is more fully referred to below.

Shear.

The behaviour of light alloys in shear is particularly important in connexion with their use for structural purposes, and especially in the form of rivets. Some misgivings in regard to this matter appear to have existed both in this country and abroad, since in the earlier rigid airship structures the rivets were made either of pure aluminium or of a very soft alloy. More recently the suitability of the harder alloys for this use has been recognized.

Shear tests on alloy "A" (3/20) have been made both on sheet and on thin rod or wire in the form of rivets. The tests on sheet were made (in the Engineering Department of the Laboratory) by punching out small disks 9·5 mm. in diameter. The results are shown in Table 19.

The results in Table 19 are in each case the mean of six closely-agreeing determinations. The material used was a typical sample of alloy "A" in its best condition, tensile tests having shown an ultimate stress of 27·7 tons per square inch and an elongation of 22 per cent on 2 inches. It will be seen, therefore, that the shear-strength of the material is appreciably lower than the tensile strength but even in shear it is still relatively very strong.

TABLE 19.

Shear Tests on sheets of Alloy "A."

Thickness of sheet.	Ultimate Shear Stress.
	Tons per square inch.
0·048 inch	21·9
C·061 ,,	23·3

Tests on rivets were made by breaking actual riveted joints made with either a single rivet or with two rivets. The rivets used were made from cold-drawn wire which was annealed at 250° C. before use. The rivets were 0·063 inch in diameter, and gave ultimate shear stresses of 17·3 tons per square inch in tests on two rivets and of 15·8 tons per square inch in a test on a single-rivet joint. These somewhat lower values, as compared with the results on sheet shown in Table 19, are accounted for by the fact that the material of the rivets had been annealed so as to render it as soft as possible. Even in this condition, however, the rivets of alloy "A" (3/20) are much stronger than the pure aluminium or soft-alloy rivets formerly employed, which gave shear-strengths, when tested in riveted joints, of only 7 to 8 tons per square inch.

Fatigue.

Some fatigue tests on the preliminary material of the ternary alloys described above in Section III (a) were carried out on the special machine for the application of alternating load in direct compression and tension which was at that time used for this purpose in the Engineering Department of the Laboratory. At the time when rolled material of alloy " A " (3/20) became available for tests of this kind, the Engineering Department had adopted the Wöhler type of testing-machine for fatigue determination. Fatigue data on rolled-bars of alloy " A " are therefore given as results of the Wöhler tests. These determinations, taken both at the ordinary temperature and 150° C., are embodied in Table 45 (page 150). This Table contains data for comparison of a whole series of alloys among which the alloy " A " (3/20) is included. This Table is more fully discussed below (Section III, e).

Notched-Bar Impact Tests.

Since it has been recognized for some time, both by the present Authors and others that the notched-bar impact test, in the present state of our knowledge, cannot be satisfactorily employed for the purpose of comparing the merits of materials of different composition and constitution, this test has been mainly employed in the present research for the purpose of studying the effect of different modes of treatment on the same alloy. A series of values fo a number of different alloys will be found in the present Report, but it is not considered that the various alloys can be reasonably compared on the basis of these figures.

Notched-bar impact tests have been made on different forms of the alloy " A " (3/20), and various forms of test-piece have been used, corresponding partly to the variations in the form of the material, such as sheet or rod, and depending also in part upon the particular type of test-piece which had been adopted or was being tried by the Laboratory at the period when the tests were made.

An early series of tests, carried out in April 1916, were made on test-pieces of the Charpy International type, having a cross-section

of 10 mm. square with a length of 53·3 mm., and a distance
between supports of 40 mm. The notch was 5 mm. deep, having
a radius at the root of ⅔ mm. The energy of blow used was
2·7 kilogramme-metres. These tests were made on specimens cut
from 1⅝ inch rod, rolled from previously forged billets.

Three tests gave the following results :—

Energy absorbed in Fracture	Angle of Fracture
(Kgm/M per square cm.)	(Degrees).
2·01 ⎫	11·5
2·28 ⎬ mean = 2·15	17·5
2·16 ⎭	12·0

The next series of notched-bar impact tests on this alloy were
made in September 1918, for the purpose of ascertaining the effect
on impact resistance produced by the cross rolling of alloys during
the breaking down process. The tests were accordingly made on
material taken from rolled plate ¼ inch thick, which had been
produced by hot rolling after forging from slabs 1½ inch thick. A
portion of this material had been rolled down without crossing,
while the remaining portion had been cross rolled. The test-pieces
used from this material measured 5 mm. square in cross section
and had a V notch 1 mm. deep, with an angle of 90° and a root
radius of ⅔ mm. The length of the test-pieces was 27 mm., and the
distance between supports 20 mm. These were tested on a
small impact testing-machine of the Charpy type, which had been
specially constructed in the Engineering Department for the purpose
of dealing with very small notched bar test-pieces. The energy of
the blow in this machine was 1·67 kilogramme-metre. The results
are given in Table 20.

It will be seen that these tests are, with one exception,
remarkably regular. It is also evident that cross rolling improves
the quality of this material quite appreciably. Not only is the
notched-bar impact figure on a transverse test-piece very nearly
doubled, but even the longitudinal test-piece is nearly 30 per cent
better in material which has been cross rolled. For the purpose of
comparison with the above figures relating to small test-pieces with

V notch, taken from rolled plate, the results of a similar test, on specimens of the same dimensions, cut from a ⅞-inch diameter rod, hot-rolled, may be quoted. This showed an absorption of energy of 0·53 Kgm/M., a figure which agrees very well with that from the

TABLE 20.

Condition of material.	Direction of test-piece.	Energy absorbed in fracture (Kgm/M.).
Cross-rolled.	Longitudinal	0·82
	,,	0·83 Mean 0·85
	,,	0·90
	Transverse	0·70
	,,	0·54 Mean 0·65
	,,	0·70
Not cross-rolled.	Longitudinal	0·64
	,,	0·51 Mean 0·59
	,,	0·63
	Transverse	0·33
	,,	0·34 Mean 0·35
	,,	0·38

corresponding longitudinal test-piece from plate metal which had also been rolled in one direction only.

Influence of Impurities on Alloy "A" (3/20).

The influence of impurities on a given alloy must be considered from two points of view, according to their effect (a) upon the manufacture of the alloy, and (b) on the physical and chemical properties of the finished material.

With regard to the effect of impurities on the production of the alloy "A" (3/20), some reference has already been made above

(*see* page 67) to the failure during rolling which resulted from incorporation in the alloy or in other alloys of a similar type, of various impurities. In those experiments, in fact, the only billet which was successfully rolled was one containing 1 per cent of iron. Subsequent experience, gained throughout the long course of experiments on the production of this alloy, has strongly confirmed the conclusion that the presence of impurities or small intentional additions of other metals immediately increases the difficulty in forging and rolling. This was experienced to a very striking extent when a systematic endeavour was made to incorporate in the alloy $\frac{1}{2}$ per cent of manganese. The addition of a proportion of $\frac{1}{2}$ per cent of this element was considered desirable on account of the beneficial influence which it undoubtedly exerts in lessening the risk from spontaneous or so-called " season-cracking," which is liable to occur in the case of alloy " A " if it has been wrongly treated (*see* below, pages 177 to 186).

The presence of manganese, or of iron or similar added elements, whether as impurities or as intentional additions, causes brittleness in the alloy, particularly in the early stages of working, during forging and breaking down. This is no doubt due to the occurrence in the alloys of hard, brittle constituents, generally inter-metallic compounds, which have only a very limited degree of solubility in solid aluminium, and therefore make their presence felt as separate constituents, even when the amount of the added elements is quite small. In addition to the elements already named, this applies particularly to tungsten, molybdenum and chromium. Elements, on the other hand, which are capable of entering into solid solution in aluminium, or enter into the formation of a ternary compound with aluminium and one of the other metals present, behave in a very different manner. In some cases these also materially increase the hardness or brittleness of the alloy. In other cases, on the other hand, their presence or addition actually renders the alloy more ductile and easier to forge and roll. This applies to a certain limited extent to silicon, and, to a greater extent, to nickel. Further reference to the action of nickel in this respect will be made below (page 184).

As regards the effect of various added elements or impurities on the tensile properties of alloy " A " (3,20), a certain amount of data are available and are recorded below in Table 21, the chemical composition, in most cases from analysis, of the alloys represented

TABLE 21.

Effect of Added Elements on Alloy " A " (3/20).

Distin-guishing number.	Nature of material.	Elements added.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation per cent on 2 inches.
W 685.	Hot-rolled sheet 0·045 inch thick	Iron and Silicon	19·7	28·2	18
14R	Hot-rolled rod ⅞-inch diameter	Iron	18·1	29·5	18
W560B	Hot-rolled sheet 0·069 inch thick	Silicon	17·6	28·3	16
W326	Hot-rolled sheet 0·035 inch thick	Manganese	16·4	28·7	21
W702	Hot-rolled sheet 0·062 inch thick	Manganese	18·6	28·4	21
J	Hot-rolled sheet 0·088 inch thick	Magne-sium and Silicon	21·6	24·7	15
W763A	Hot-rolled sheet 0·050 inch thick	Nickel	24·4	28·0	13

by the various numbers in Table 21 being given under corresponding numbers in Table 22.

It will be seen that in the majority of cases the presence of the added element renders the alloy slightly harder and stronger but correspondingly less ductile. The effect on the ductility of the

alloy in the earlier stages of forging and rolling, however, is much more marked than the results of the tests on the finished material would indicate.

Strength at Low Temperatures.

Together with a number of other materials, alloy " A " (C/20) has been subjected to a series of tensile tests at very low temperatures

TABLE 22.

Composition (by Analysis) of Alloys of Table 21.

Distinguishing Number.	Copper.	Zinc.	Iron.	Silicon.	Manganese.	Magnesium.	Nickel.
W685 .	3·10	20·28	0·59	0·49	—	—	—
14R . .	2·85	19·59	1·28	0·20 } 0·71 }	—	—	—
W560B .	—	—	—	..	—	—	—
W326 .	2·86	21·00	—	—	0·53	—	—
W702 .	2·27	20·02	—	—	0·41	—	—
J. . .	3·08	20·42	—	0·68	—	0·43	—
W763A*	2·50	20·0	—	—	- -	—	0·5

*Nominal composition—not analysed.

which were undertaken in view of the possibility that alloys employed for aircraft construction might be exposed to excessively low temperatures for relatively long periods. These tests and their detailed results are described, together with those for other alloys, in Section III (e) (page 142). Here it need only be said that the results clearly indicate that even prolonged exposure to the temperatures either of liquid air (- 185° C.) or of melting carbon-dioxide snow (- 80° C.) produces no deleterious effects in alloy " A." There is no evidence of any tendency towards any " embrittling " action of such intense cold.

III(z). *Development of Alloys " E " and " Y."*

It has already been indicated that the study and development of the ternary alloys of copper and zinc with aluminium was undertaken partly with a view to the subsequent utilization of the alloys arrived at as a basis for further advance. In the case of the alloys of manganese and copper with aluminium, which are described in the later portion of the Ninth Report to the Alloys Research Committee, such a further advance had already been effected. This was first done by Wilm, who applied the hardening effect produced by the addition of magnesium to aluminium alloys, to these materials, which at that time constituted the best wrought

TABLE 23.

	1	2	3	4	5	6	7	8	9	10
Copper . . .	3	3	3	3	0	0	0	0	3	3
Zinc	5	10	15	15	15	20	15	20	15	20
Magnesium. .	0·5	0·5	0·5	0·5	0 5	0·5	0·5	0·25	0·25	0·5
Manganese . .	--		—	1·0	--	—	--		—	—
Silicon . . .	Not exceeding 0·20.									
Iron	Not exceeding 0 20.									
Aluminium. .	Remainder.									

aluminium alloys available. Since the aluminium-zinc-copper alloys, whose development has been described in the previous sections of the present Report, constitute so considerable an advance from the point of view of tensile strength on the manganese-copper-aluminium alloys described in the Ninth Report, it was hoped that the application of the magnesium hardening effect to these alloys would bring about a corresponding further advance. Efforts were accordingly made, first to secure the incorporation of magnesium in these alloys, and in the second place, to study the whole process of hardening in the presence

of magnesium with a view to arriving at a better understanding
of the process and of the manner in which it could be applied to
a new group of alloys. This latter part of the work is described in
a separate section of the present Report (*see* Section V(*d*) (page 235).

The first attempts at rolling alloys containing magnesium were
made at the Milton Works of the British Aluminium Co., in November
of 1911, and again at the same works in 1913. The compositions
of the alloys used in those experiments are stated in Table 23.

Alloys having the composition indicated in columns 1-6 of the

TABLE 24.

Alloy.	Material.	Condition.	Yield. Tons per sq. inch.	Ultimate. Tons per sq. inch.	Elongation per cent on 2 inches.
7	⅞ in. rod	Hot rolled	19·9	25·3	21
8	,, ,,	,, ,,	21·7	25·7	19
7	⅞-in. rod	Quenched from 500°	12·4	20·2	29
8	,, ,,		15·5	23·1	29
7	⅞-in. rod	Quenched from 500° and aged at least one week	15·4	22·5	27
			18·9	24·9	18
15 per cent zinc	⅞-in. rod	Hot rolled	8·5	16·5	32
20 ,, ,, ,,	,, ,,	,, ,,	12·4	21·4	26

Table were prepared in the form of 3-inch chill cast billets, two
billets of each composition being tried, one for rolling at a relatively
high temperature (440° C. to 390° C.), and the other at a lower
temperature (325° C. to 275° C.). At both temperatures, however,
all the billets of these compositions broke up badly in the
rolls, generally at the first pass. Of the remaining alloys, those
corresponding to columns 7 and 8 were successfully broken down at
436° C. and 418° C. respectively and were ultimately rolled to
⅞ inch diameter rod. Tensile tests on the resulting material, both

in the hot-rolled condition and after quenching and ageing are given in Table 21. For purposes of comparison similar data for alloys of the same zinc content, but without magnesium, are included in this Table.

It will be seen from the above Table, that both alloys in the hot-rolled condition give results decidedly higher than the pure aluminium-zinc alloys free from magnesium. After quenching, the alloys containing magnesium were very much softer than in the hot-rolled condition. A certain amount of hardening occurred on ageing, but this was not sufficient to restore the alloys to the strength which they possessed after hot rolling alone. In view of other developments, the study of zinc-aluminium alloys containing magnesium has not been carried further.

After the installation of the experimental rolling-mill at the National Physical Laboratory, the study of the behaviour of ternary alloys, to which magnesium had been added, was carried on concurrently with that of the ternary alloys without magnesium, which has been described above. The alloys used had the composition indicated in Table 25.

TABLE 25.

—	9	10	11	12	13
Copper	3	3	3	2·5	2·5
Zinc . . .	15	20	20	13	13
Magnesium . .	0·25	0·5	0·25	0·5	0·5
Silicon* . . .	—	—	—	—	1·0
Iron*			—	—	
Aluminium . .			Remainder.		

* Not exceeding 0·2 except where otherwise stated.

The first attempts to roll billets of these alloys without previous forging led to consistent failure When forging was first attempted, it was found that the presence of magnesium in the alloys very

considerably increased the difficulty of forging, as there was a
greater liability to crack. In view of the difficulty of forging
these alloys, attention was turned at this stage to the possibility
of bringing them into workable condition by use of the extrusion
process.   The work on extrusion is more fully dealt with
in a separate section of the present Report, but it is desirable
to record in this place that alloys of compositions 9 and 12, Table
25 were more or less successfully extruded, and in the extruded
state proved readily amenable to hot-rolling.   In addition to the
above, one very considerably harder alloy, subsequently known
as alloy " E," and having the nominal composition : —

					Per cent.	
Copper	.	.	.	.	2·5	
Zinc .	.	.	.	.	20·0	
Magnesium .	.	.	.	0·5	
Manganese .	.	.	.	0·5	
Iron. .	.	.	.	.	0·20	} not exceeding,
Silicon	.	;	.	.	0·20	

was also successfully extruded, although a part of the extruded
bar was seriously cracked. After rolling sound portions of this
material down to a diameter of $\frac{3}{4}$ inch, it gave tensile test results
as shown in the Table 26.

TABLE 26.

Primitive El. Limit. Tons per sq. inch.	Yield stress. Tons per sq. inch.	Ultimate stress. Tons per sq. inch.	Elongation per cent on 2 inches.
22·1	32·9	37·8	12
18·9	32·1	36·8	11

The elasticity curves and the load-extension diagrams for the two
specimens, from which the results given in Table 26 were obtained,
are shewn in the graphs of Figs. 60 and 61.

.The above results constitute a considerable advance on any
light alloys previously produced, so far as tensile properties are

Fig. 60.—*Alloy " E."* ⅞-*inch Rod rolled from 1½ inch, Extruded.*
Quenched 350° C. and Aged.
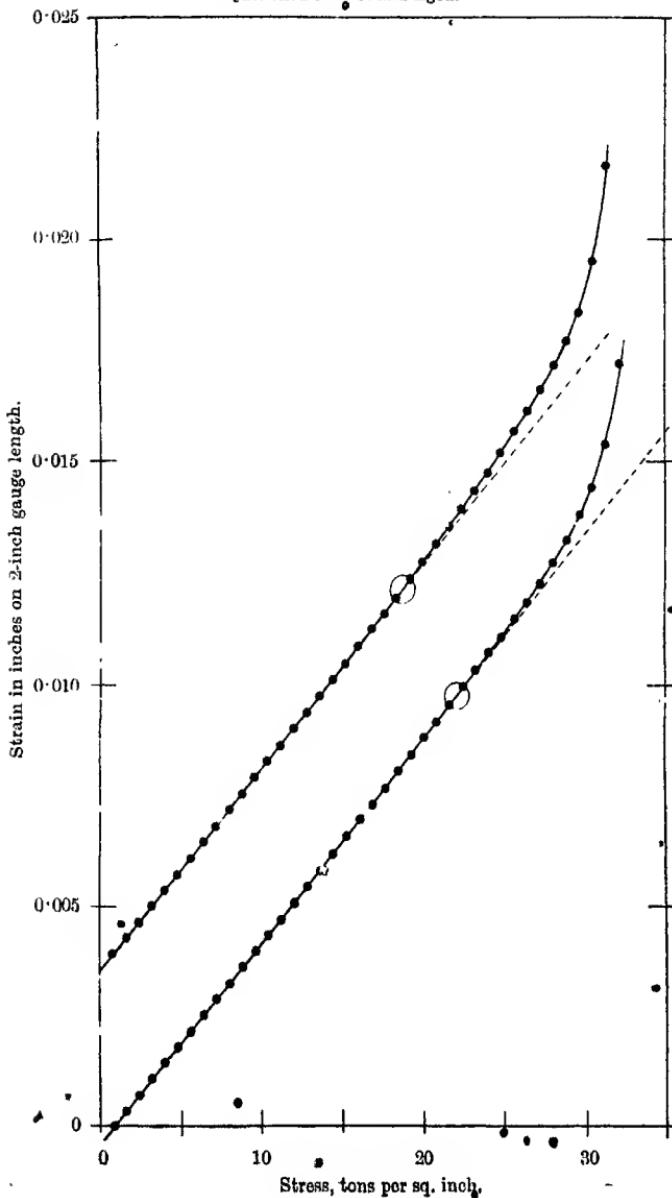
Fig. 61.—*Load-Extension Diagrams of Extruded-Rolled Rods of Alloy "E."*

Strain in inches on 2-inch gauge length.

0   0·05   0·10   0·15   0·20   0·25   0·30

Stress, tons per sq. inch.

5   10   15   20   25   30   35
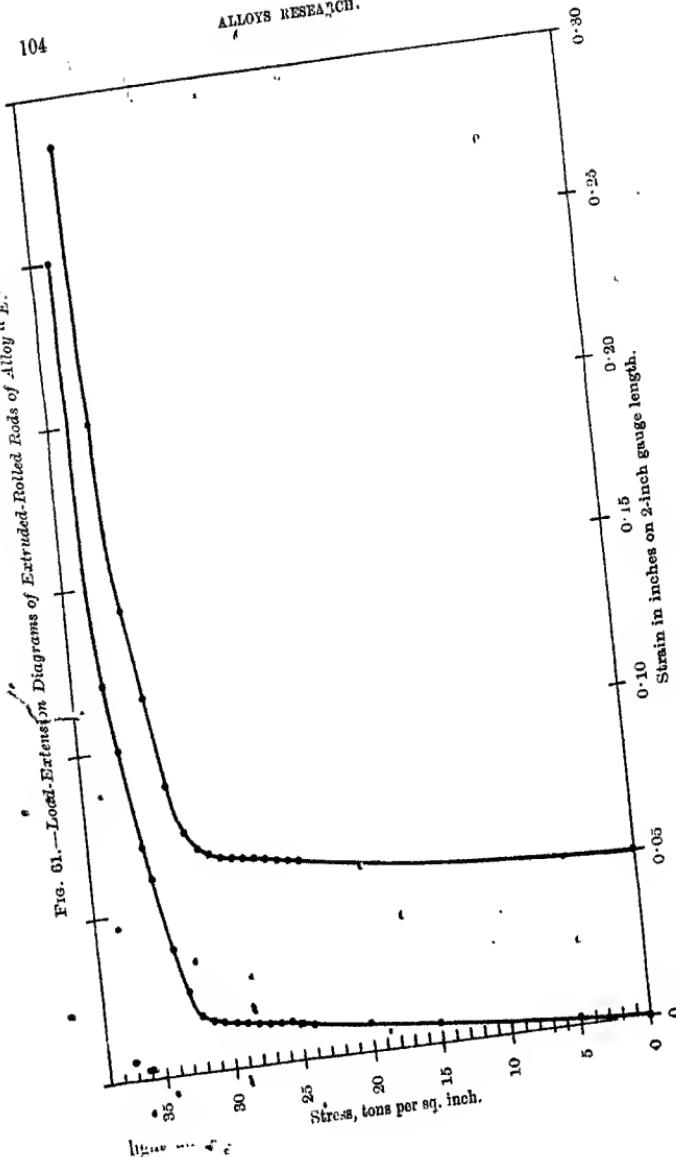
concerned. Special attention was therefore devoted to the further development of alloys of this type, and particularly to their production in the form of sheet. For this purpose, the extrusion process which proved itself very useful for the production of alloys in the form of rods and bars, was much less suitable, and for this and other

FIG. 62.

Sheet (18 Gauge).   Compositions 12 and 13.
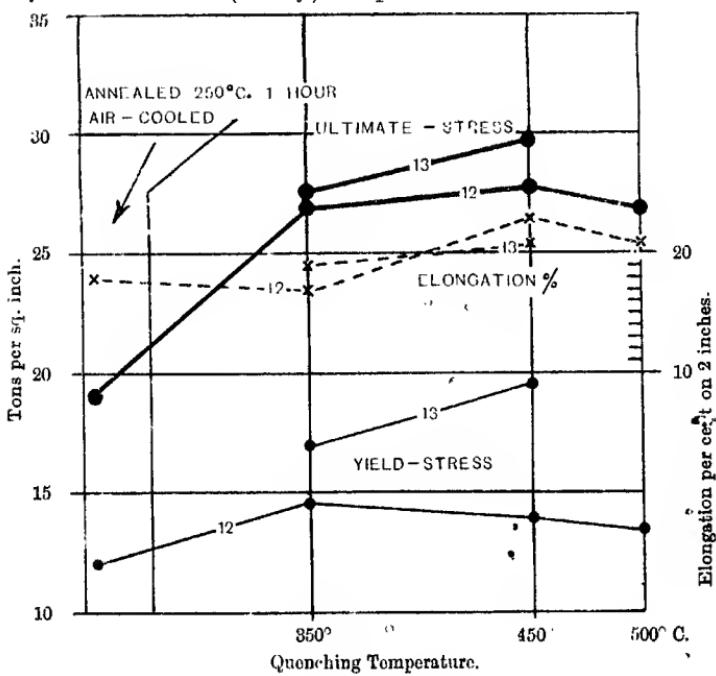
reasons it became necessary to continue the study of methods of producing these alloys from cast slabs by rolling.

Concurrently with the extrusion experiments described above, further efforts were made at forging and rolling slabs $\frac{3}{4}$ inch thick. In the case of compositions 12 and 13, Table 25, some success was obtained. The results of tensile tests on 18 gauge sheet of those

compositions (12 and 13) after annealing at 250° C., and after
quenching from 350° C., 450° C. and 500° C., followed by ageing
from five to ten days; are shown in the curves of Fig. 62. These
results indicate marked age hardening, the tensile strength being
increased by as much as 50 per cent above that of the annealed
material and 30 per cent above that of the material as hot rolled.
The alloy corresponding to column 13, containing intentionally
added silicon, proved to be more readily rolled and was practically
free from edge cracking.

The experience which had been gained at this stage of the
research, in the production of rod and sheet of the alloy "A" (3/20),
suggested that possibly the difficulties met with in the attempts
to forge and roll the harder alloys containing magnesium might be
traceable to defects in the billets and castings. Special attention
was therefore devoted to the method of making and casting these
alloys.

Method of Making and Casting Alloys Containing
Magnesium.

In making these alloys, magnesium had hitherto been added
in the form of an alloy of magnesium and aluminium containing
20 per cent of the former metal. Slabs made in this way, however,
were found to be seriously defective when examined microscopically.
This appeared to be due to the inclusion of oxide, coupled with
unsoundness, and it seemed probable that a large part of the
difficulty from cracking during both forging and rolling might be
attributable to these defects. Micro-sections of the magnesium-
aluminium alloy showed similar defects, and its use was therefore
abandoned. The practice was adopted of adding the magnesium
in the form of the pure metal, cut into one or two pieces of fair
size. These were added to the molten alloy just before pouring.
Slabs made in this way were found to be practically free from
defects and gave very much improved results on working. The
magnesium was introduced into the molten alloy by the device
which has already been described in connexion with the production
of casting alloys containing magnesium (see Fig. 20, page 33).

Forging and Rolling of the Alloy.

Preliminary tests on strips cut from $\frac{3}{4}$-inch slabs were made by forging at various temperatures. It was found that the lowest

Fig. 63.—*Forging Tests. Alloy "F"*
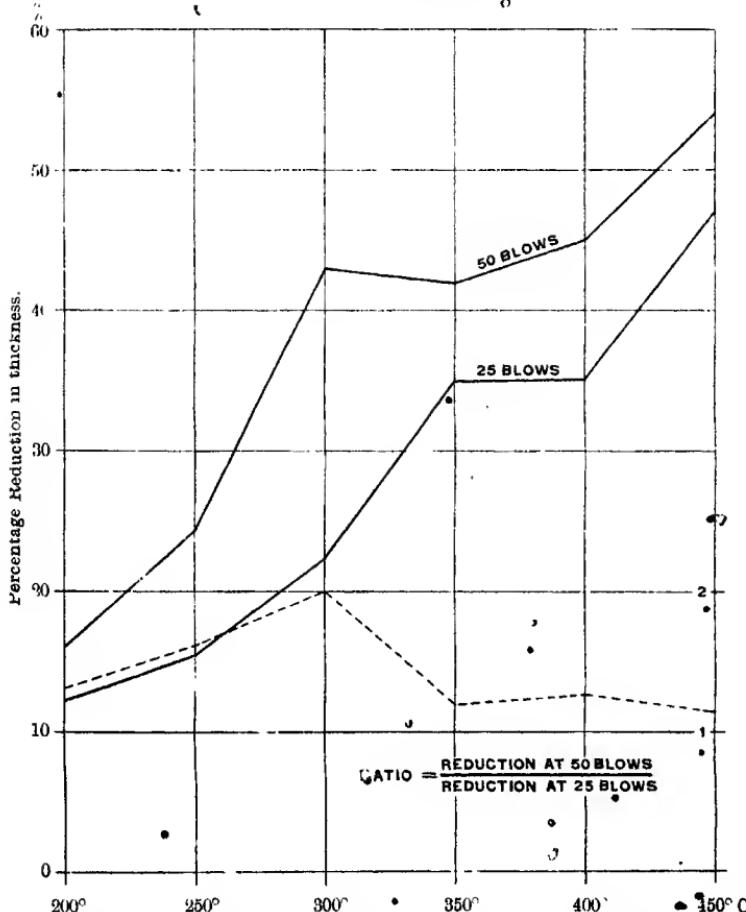
temperature, 350° C., was the most satisfactory, and that as much as 30 per cent reduction could be obtained without cracking.

Subsequently it was found preferable to use a still lower working temperature, 300° C.

This is illustrated by a series of forging tests made on strips cut from larger chill cast slabs measuring 7 inches by 14 inches by 1½ inch thick, the strips themselves being 5 inches long by 1½ inches square section. Forging was confined to an area at one end which could be covered by a single blow. The results are shown in Fig. 63, showing the reduction in thickness produced by twenty-five and

TABLE 27.

Tensile Tests. Alloy "E." 18 Gauge Sheet.

Condition.	Yield stress. Tons per sq. inch.	Ultimate stress. Tons per sq. inch.	Elongation per cent on 2 inches.
As hot-rolled . .	26·3	31·4	15
	24·8	32·6	16
Annealed 250° C. . .	14·5	25·3	20
Quenched 250° C. and aged	13·7	23·3	22
Quenched 350° C. and aged	27·3	36·5	10
Quenched 400° C. and aged	27·8	38·7	9
	30·5	39·1	11

fifty blows respectively at various temperatures. At 450° C. it appears that the alloy is hardened so much by twenty-five blows that a further twenty-five blows produce little more reduction, while at 300° C. such an action does not occur.

Ultimately it was found that by using a blow distinctly lighter than that employed in the case of the alloy "A" (3/20), slabs could be reduced in thickness of from 5 to 10 per cent in one stage. After this preliminary forging treatment, it was found possible to break the slabs down in the rolls at a similar temperature (300–320° C.) and to effect a 50 per cent reduction in four passes. For further rolling of blanks to thinner sheets, a higher temperature, 400° C., was adopted.

With this alloy, edge-cracking during breaking down was never completely eliminated, as had been done in the case of the alloy "A" (3/20), so that a considerable percentage of scrap could not be avoided. Once the material had been reduced 50 per cent in thickness and had been trimmed to remove cracked and defective portions little further trouble with cracking was experienced.

Results typical of tensile tests of this alloy "(E)" after rolling to a thickness of 0·05 inch, and after receiving the various forms of heat-treatment indicated, are given in Table 27.

It will be seen that this material attains a tensile strength of from 38–39 tons per square inch when fully hardened and aged. After quenching at 350° and 400° C., however, the ductility, as shown in the above Table, appeared unsatisfactory, and further work was undertaken with a view to effecting an improvement in this property without undue sacrifice of tensile strength. The composition of the alloy was modified in two ways, (1) by reducing the magnesium content, (2) by reducing the magnesium content and adding silicon. The compositions tried were as follows : —

					1	2
Magnesium	.	.	.	.	0·25	0·25
Manganese	.	.	.	.	0·5	0·5
Silicon	.	.	.	.	0·5	1·0
Copper	.	.	.	.	2·5	2·5
Zinc	.	.	.	.	20·0	20·0
Aluminium	.	.	.	.	Remainder.	

The results obtained with material of these two compositions, treated otherwise in exactly the same manner as that described above, are given in Table 28.

It will be seen that both modifications show a material improvement in regard to elongation, with only a comparatively small reduction in tensile strength.

Subsequent to the successful production of the alloy " E " at the
Laboratory, its manufacture on an industrial scale was studied and
developed at the Royal Airship Works at Cardington, at that time
under Messrs. Short Bros. Efforts were there concentrated mainly
upon the reduction in the percentage of scrap, which, owing to the

TABLE 28.

Condition.	Yield stress. Tons per sq. inch.		Ultimate stress. Tons per sq. inch.		Elongation per cent on 2 inches.	
	1	2	1	2	1	2
As hot-rolled . . {	27·1	24·8	32·2	30·8	18	15
	26·9	26·9	31·9	31·6	15	15
Quenched 350° and aged {	26·6	22·2	36·3	33·7	15	12
	28·4	20·6	36·1	33·8	14	12
Quenched 400° and aged {	28·6	26·0	38·2	35·4	13	15
	28·4	28·4	36·0	35·3	14	16

occurrence of edge cracking, was still somewhat serious. Modifications
in the procedure of manufacture, and also some modifications of
composition suggested by the present Authors were tried. The
modifications of procedure consisted mainly in (1) machining of the
surfaces of all cast slabs, and (2) modifications of composition

(1) *Machining of Slabs.* –It is probable that if entirely perfect
slabs free from surface defects could be produced, machining would
not be necessary and would indeed offer no advantages. In actual
industrial practice, however, it seems that machining of the surface
is probably cheaper and more effective than the rejection of all
slabs showing minor defects. A further modification of treatment
was introduced after machining, by submitting the slabs to a
preliminary annealing before forging. After this preliminary
annealing, and using the larger slabs and heavier hammer installed
at those works, it was found possible to give the slabs a reduction

in forging of 30 per cent in thickness. After this treatment the rolling presented fewer difficulties, and the amount of scrap was very largely diminished. In addition, it was found, much to the Authors' gratification, that the more drastic treatment, which could be given to the material in a larger plant, led to an improvement in the ductility of the ultimate product. This will be seen by comparing the results given in the summary Table 40, page 144.

(2) *Modifications of Composition.*—In order to obtain a somewhat softer alloy, two modifications of composition were suggested by the Authors and were tried at Cardington. The first of these, known as alloy "F," consisted in the addition of a small amount of silicon (such as to bring the total silicon in the alloy up to approximately 0·75 per cent). As will be seen from the summary Table 40, page 144, this effected a certain amount of softening, reducing the average tensile strength to 35 tons per square inch as compared with 39 tons per square inch of the alloy "E," with a corresponding increase in elongation to 18 per cent as compared with 14 per cent of alloy "E." A further modification, leading to the alloy known as "G," was made by reducing the percentage of all the added elements except copper, including zinc, manganese and magnesium. The composition aimed at was as follows :—

Copper	.	.	.	.	.	.	2·5
Zinc	.	.	.	.	.	.	1·80
Magnesium	.	.	.	.	.	.	0·35
Manganese	.	.	.	.	.	.	0·35
Silicon	.	.	.	.	.	.	0·80
Aluminium	.	.	.	.	.	remainder

This change in composition produced very little diminution in tensile strength as compared with alloy "F," but a considerable increase in elongation was obtained. This also will be seen on reference to the summary Table 40, page 144.

Development of Alloy " Y."

The work on casting alloys for use at high temperatures, described in Section II b of the present Report, had indicated

the remarkable properties, particularly at high temperatures, which were found in the alloys of aluminium with copper and nickel and still more in the alloys with copper, nickel and magnesium. Not only was it found that these alloys maintained their strength at a high temperature to a greater degree than any others which had been tested, but they combined with this property the development of a degree of ductility, remarkable in cast material, in that range of temperature which is usually employed for the rolling of aluminium alloys, that is, the range from 300° to 450° C. This ductility, which exists even in alloys containing relatively high percentages of copper and nickel, suggested the possibility that these materials might be suitable for forging and rolling, and it was further hoped that in the wrought condition, particularly after heat treatment, they might still display the power of retaining their strength at high temperatures to a special degree. This was regarded as of particular importance because alloys of the type of " A " (3/20), and of the type of " E " and its modifications, however useful and promising at ordinary temperatures, were remarkably deficient when any attempt was made to use them at elevated temperatures. For such purposes as connecting-rods of aero engines, for example, it was evident that the alloys containing notable proportions of zinc were entirely unsuitable. This latter class of alloy, also, had shown certain definite limitations arising from such matters as " season cracking " and corrosion (see below, page 177), which rendered it desirable to look in other directions for alloys which should as far as possible combine the remarkable strength of the zinc alloys with freedom from these limitations. It is in this direction particularly that alloys of the " Y " type, containing copper, nickel and magnesium, have been developed and studied.

Exploratory Work.

prel.a
anneali
at those we   eries of exploratory meltings were made, the alloys being
     the form of rods 1⅛ inch diameter and 16 inches long in
     moulds. The (nominal) compositions of these alloys

together with their behaviour on rolling, are shown in Table 29.

The reduction aimed at in rolling was from $1\frac{1}{8}$ inch to $\frac{5}{8}$ inch diameter. It will be seen that with the exception of one or two bars containing a high percentage of magnesium (3 per cent) good rolled rod was obtained from all the alloys. Special attention may be drawn to the fact that the alloy E1, Table 29, containing copper 9 per cent and nickel 3 per cent was satisfactorily rolled. In the Eighth Report to the Alloys Research Committee, Carpenter and Edwards state that alloys of aluminium with copper alone, containing more than 4 per cent of copper, can only be rolled, if at all, with very great difficulty. With the facilities and knowledge now available, this limit could probably be extended somewhat, and alloys of the " D " type (resembling Duralumin) have been rolled containing copper up to 6 per cent. There can be little doubt, however, that an alloy containing 9 per cent of copper alone could not be rolled satisfactorily if at all. We have here, therefore, the very remarkable fact that the addition of 3 per cent of nickel, which by itself would act as a hardening agent, markedly increases the ductility of an alloy containing 9 per cent of copper, particularly at high temperatures. Further reference to this influence of nickel will be made below.

A series of tensile tests, which could only be regarded as preliminary, were made on the rolled-rod obtained from the above series of alloys. These tests gave low values, probably because the amount of reduction which had been possible in the rolling of these small chill cast ingots had not been sufficient fully to develop the strength of the material. On the other hand, these tests showed that the rolled alloys at temperatures up to $200°$ C. maintain their strength very well, particularly those containing copper, nickel and magnesium. The loss of tensile strength at $200°$ C., in the case of the alloy containing copper 4 per cent, nickel 2 per cent, magnesium 1·5 per cent (subsequently known as the " Y " alloy), did not amount to more than 10 per cent of the strength at the ordinary temperature. This may be compared with a loss of about 50 per cent for hot-rolled rod of alloy " A " (3/20).

TABLE 29.

Exploratory Alloys. Compositions and Behaviour on Rolling. Billets 1½-inch diameter, 16 inches long.

Alloy	Composition			Behaviour on Rolling.
	Cu.	Ni.	Mg.	
E1	9·0	3·0	—	Good bar.
E2 (1)	4·0	2·0	1·5	Twisted and cracked badly one end.
E2 (2)	4·0	2·0	1·5	Split in half at first pass, scrapped.
E2 (3)	4·0	2·0	1·5	Good bar.
E2 (4)	4·0	2·0	1·5	Good bar.
E3	4·0	2·0	1·0	Good bar, re-heated once during rolling.
E4	4·0	4·0	1·0	Fair bar, cracked one end.
E5	4·0	2·0	3·0	Split in half at last pass, scrapped.
E5	4·0	2·0	3·0	Cracked first pass, scrapped.
E6	2·0	1·5	1·0	Good bar.
E7	—	6·0	—	Good bar.
E8	—	6·0	1·5	Good bar.
E9	—	6·0	3·0	Cracked badly at second pass.
E9	—	6·0	3·0	Good bar, re-heated after second pass.

Larger Billets.

The production, particularly of the " Y " alloy, in the form of rods hot-rolled from 3 inch diameter chill cast billets was next studied. At the same time, in order to ascertain the effect of modifications of composition upon the strength of the alloy, some billets of modified composition were cast and rolled. For purposes of comparison, a series of alloys of the " D " type (resembling Duralumin in containing about 0·5 per cent of manganese, with 0·5 magnesium) were cast ; in one of these the copper content was increased to 6 per cent and in another an addition of 2 per cent of nickel was made. These were rolled and treated together with alloys of the " Y " series. The compositions of the alloys of this series, as found by analysis, are given in Table 30 together with the results of breaking down in the rolls from 3 inch diameter to ⅞ inch diameter hot-rolled rods. The alloys marked with an asterisk in the Table were forged before rolling, but the first two, of the " Y " type proper were rolled without preliminary forging.

It will be seen from the Table, that no great difficulty was experienced in rolling these alloys except in the case of the " D " type alloy containing 6 per cent of copper, which split somewhat at one end. The " Y " alloy, both in this series of tests and the previous series from 1⅛ inch diameter chill billets, proved capable of being rolled without preliminary forging. It was subsequently found, however, particularly in the case of rolling of sheet, that it is a distinct advantage to carry out a certain amount of forging before breaking down in the rolls.

Tensile Tests.

Tensile tests have been made on the rod, ⅞ inch diameter, obtained from each of the alloys. The tests have been made at normal temperatures and at 100° C., 150° C. and 200° C. In all cases these materials had been previously heat-treated, by being quenched from a temperature of 480° C. in cold water and aged for at least four days. The results of the tests are represented graphically in Figs. 64 (page 118) and 65 (page 119), the former referring to the alloys of the " Y " type, and the latter to alloys of the " D " type. The letters and numbers

TABLE 30.—3 inch dia. Billets to $\frac{5}{8}$ inch dia. Rod.

Alloy	Composition by Analysis.					Behaviour on Rolling.
	Cu.	Mn.	Ni.	Mg.	Si.	
	Per cent	Per cent	Per cent	Per cent	Per cent	
Y Type.						
P1.	4·22	0·4	2·02	0·52	0·79	Good bar, rolled without trouble.
P2.	4·08	—	2·03	1·6	—	,, ,, ,, ,,
P3*	4·17	—	1·79	1·75	0·69	{ Split a little one end at fourth pass, other end also split a little at next pass.
P4*	4·35	0·4	1·77	1·89	—	{ Split a little one end at fifth pass, reversed and finished well.
P5*	5·05	—	2·51	1·75	—	Split a little one end, finished well.
P6*	6·03	—	3·09	2·00	—	{ Split badly both ends, was cropped and finished well.
P7*	2·0	—	1·45	0·95	0·75	
Duralumin Type.						
D4*	4·0	0·5	—	0·5	0·75	Good bar, rolled without trouble.
D5*	5·08	0·4	—	0·85	0·89	,, ,, ,, ,,
D6*	6·25	0·47	—	0·78	1·17	{ Split down half its length at third pass, reversed, no further trouble.

* Forged before rolling.

of these Figures are the same as the identification letters and numbers of the alloys in Table 30. In Fig. 64 a graph representing the tensile properties at various temperatures of alloy "A" (3/20) has been included for comparison. Examination of the graphs shows that on the whole, the alloy "P2" (now known as alloy Y"') gives the best results at ordinary and high temperatures. The modifications of composition tried in this series of experiments do not appear to give any advantage.

It will be seen from Fig. 64, that at a temperature of 200° C. the alloy " Y " (P2) has a tensile strength of about $19\frac{1}{2}$ tons per square inch. This corresponds to the tensile strength of this alloy as hot rolled. It would appear, therefore, that at this temperature the hardening effect produced by quenching and ageing is removed by prolonged heating. Temperatures up to 150° C., and possibly slightly higher, however, can be safely used without producing any permanent deterioration of properties. Test-pieces of this alloy have been held at 150° C. for a period of 100 hours and afterwards tested at that temperature. They showed only a negligible loss of strength compared with specimens tested after having been held at the same temperature for only thirty minutes. This matter has also been further studied by means of Brinell tests (hardness) made upon specimens held for prolonged periods at 175° C. Comparison with alloys of the " D " type (Duralumin) show that the " Y " alloy is actually superior when held at these high temperatures in retaining the hardness previously produced by heat-treatment.

Fatigue Tests at High Temperatures.

In view of the fact that it was hoped to utilize alloys of the " Y " type in the construction of connecting-rods and other moving parts of engines exposed to elevated temperatures, special importance was attached to a series of fatigue tests made at temperatures up to 150° C. It was considered that this range of temperature was sufficient for such materials, in view of the fact that even the hot end of the connecting-rod never attains anything like the full temperature of a piston. The results of these fatigue tests are given in the general Table of comparative fatigue tests, Table 45,

FIG. 64. — *Tensile Tests on Alloys, Type " Y," at Various Temperatures.*
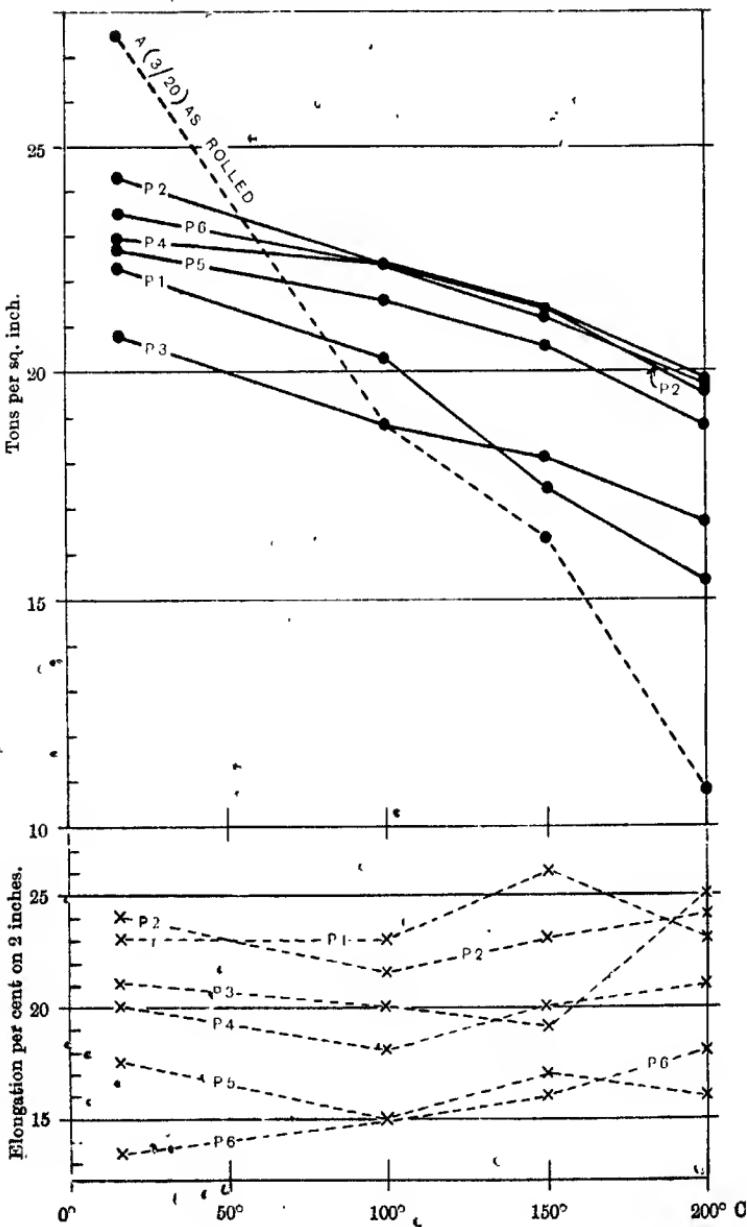
Fig. 65.

Tensile Tests on Alloys, Type "D," at Various Temperatures.
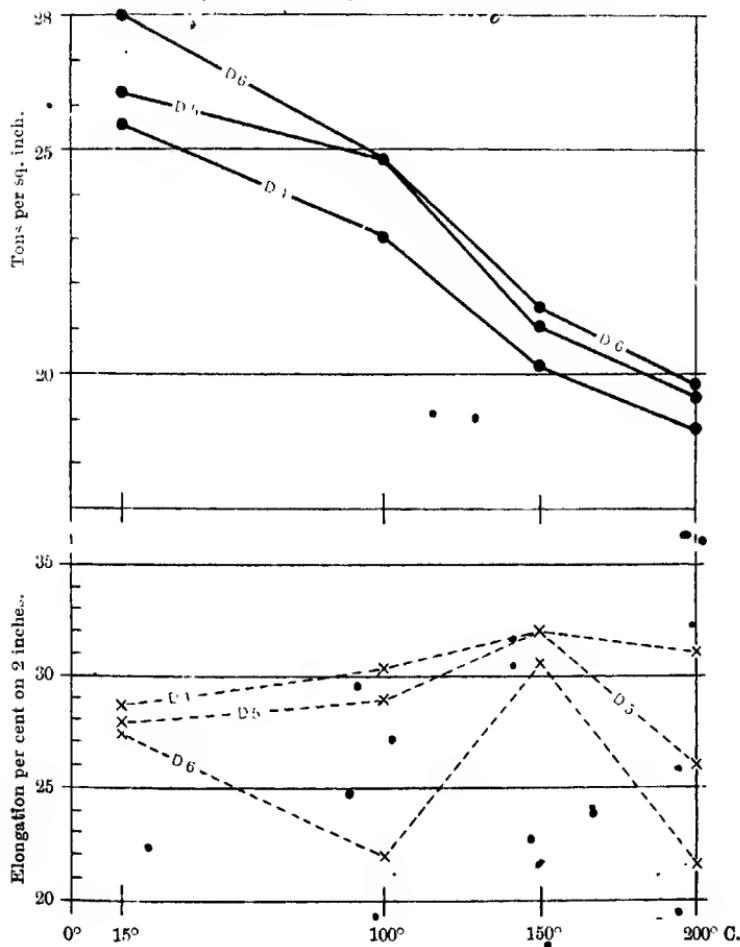
page 150. It will there be seen that the fatigue range (safe range for alternating stresses) for the "Y" alloy at 150° C. is distinctly higher than that of any other light alloy tested.

Production of Sheet.

The production of alloys of the "Y" type in the form of sheet presented comparatively little difficulty. The first experiments, were made with chill cast slabs, $\frac{3}{4}$ inch thick, and these were broken down at first at temperatures of 400° C., and later, when the results of some further forging tests had become available, at a higher temperature (180° C.). This change produced a distinct improvement in the results of working. These relatively thin slabs could be broken down with and without preliminary forging. With thicker slabs ($1\frac{1}{2}$ inch), a reduction of 10 per cent in thickness by forging was given in two heats at 240° C. This was followed by re-heating the forged blanks at 480° C. and their reduction in the rolls to a thickness of, approximately 0·75 inches in five or six passes. Typical results of tensile tests on sheet 0·05 inch thick produced in this manner, after heat-treatment, are given in Table 31.

TABLE 31.

Material.	Yield stress.	Ultimate stress.	Elongation per cent on 2 inches.
	Tons per sq. inch.	Tons per sq. inch.	
Sheet 0·05 inch thick, quenched and aged . . .	14·1	22·8	18·0
	11·8	22·1	21·0

Further Improvement of Alloys of the "Y" Type.

The advantages in regard to fatigue resistance and, particularly, in regard to resistance to corrosion which alloys of the "Y" type had been found to present, made it very desirable to study these further with a view to improving their tensile strength. Hot-rolled material

which had been quenched and aged had not given tensile strengths
beyond 25 tons per square inch, and this figure compares unfavourably
with such alloys as "A" (3/20) and still more so with alloys of the type
of "E," while even commercial Duralumin generally gives higher
figures in the heat treated and hardened condition. Efforts were
therefore made to improve these alloys in the first place by further
changes of composition. In view of the fact that the compound of
silicon and magnesium (Mg_2Si) had been found to play so large a part
in the hardening of alloys containing magnesium, it was thought that
by increasing the proportion of silicon present in these alloys, a larger
amount of compound would be produced and that a corresponding
increase in the degree of hardness might be obtained. This was
tried, but it was found not to be the case. The fuller study of these
questions of hardening is described in another section (V (d)), but
there can be little doubt that the failure to obtain increased hardening
by the addition of silicon in proportion to the amount of magnesium
present must be due to the fact that the solubility of this compound
in aluminium is restricted by the presence of relatively large amounts
of copper and nickel. The addition of silicon beyond a content of
0·35 per cent is therefore not desirable in these alloys.

A further series of alloys, in which the composition was varied
by changing in each experiment the proportion of one of the
constituents present, were also prepared and tested. These alloys
were cast in slab form, slabs $\frac{3}{4}$ inch thick, 12 inches long and 5 inches
wide being employed. The compositions of these alloys, both
nominal and as found by analysis, are given in Table 32.

The alloys of all these compositions were found to undergo
forging in a fairly satisfactory manner. On subsequent breaking
down in the rolls, generally at temperatures of 500° C they also
behaved fairly well, but as a rule there was more or less edge cracking.
This could probably be corrected by slight modifications of casting
procedure, while greater experience in rolling and treating these
alloys would no doubt lead to improved results in regard to
mechanical properties.

In order to test the effect of heat-treatment on alloys of various

composition, as shown in Table 32, the sheet produced from these alloys was treated by quenching from various temperatures, ranging from 350° C. to 450° C., and tensile tests were made on material aged after such quenching. The results of these tensile tests, corresponding to quenching from various temperatures, are shown in Table 33. It will be seen that a number of quenched strips cracked as a result of this treatment, and it would appear that

TABLE 32.

Melt No.		Composition per cent.				
		Al.	Cu.	Ni.	Mg.	Si.
W817-9	Intended	89·6	6·0	2·0	1 5	0·9
	By analysis	89·86	5·84	1·96	1·5	0·84
W850-2	Intended	89·6	4·0	4·0	1·5	0·9
	By analysis	89·9	3·87	3·88	1·5	0·84
W853-5	Intended	91·6	2·0	4·0	1·5	0·9
	By analysis	91·8	1·84	4·0	1·5	0·85
W862-4	Intended	88·6	7·0	2·0	1·5	0·9
	By analysis	88·93	6·87	1·74	1·55	0·91
W871, A.B.C.	Intended	89·4	6·0	3·0	1·0	0·6
	By analysis	89·47	6·0	2·93	1·05	0·55

quenching from temperatures as high as 450° to 550° C. in cold water is too severe a treatment for some of the harder alloys containing higher proportions of copper.

An examination of the results shown in Table 33 serves to show that none of the modifications tried have brought about any material improvement in the tensile strength of the heat treated and aged alloys. It is, of course, possible that modifications both of rolling and heat treatment might produce better results with one or other of these alloys, and it may prove ultimately to

TABLE 33.—*Tensile Tests.*

Alloy. Nominal Composition				Condition.	Yield stress. Tons per sq. inch.	Ultimate stress. Tons per sq. inch.	Elongation per cent on 2 inches.
Cu.	Ni.	Mg.	Si.				
6	2	1·5		As rolled.	11·5	21·2	7·0
				Quenched 400° C.	10·8	21·5	12·0
				,, 450° C.	10·8	23·3	15·0
				,, 500° C.	Test Strips cracked.		
7	2	1·5		As rolled.	9·6	18·0	5·0
				Quenched 400° C.	11·3	20·2	9·0
				,, 450° C.	13·5	22·5	10·0
				,, 500° C.	Strips cracked.		
6	3	1	0·6	As rolled.	15·0	22·0	6·0
				Quenched 350° C.	6·0	13·5	12·0
				,, 400° C.	9·5	17·6	12·0
				,, 450° C. (One strip cracked.)	11·0	24·0	7·0
				Quenched 500° C. (One strip cracked.)	16·0	23·4	6·0
4	4	1·5		As rolled.	7·0	20·6	6·0
				Quenched 400° C.	9·0	17·6	14·0
				,, 450° C.	8·9	17·8	14·5
				,, 500° C.	13·8	22·8	13·0
2	4	1·5		As rolled.	11·9	20·0	5·0
				Quenched 400° C.	7·2	15·4	15·0
				,, 450° C.	8·0	16·8	15·0
				,, 500° C.	13·0	23·0	14·0

NOTE.—Quenching for 450° C. to 500° C. in cold water seems too drastic for alloys with the high copper contents.

be desirable to explore the possibilities more fully. As far as the evidence goes, however, it strongly indicates that the alloy containing 4 per cent copper, 2 per cent. nickel and $1\frac{1}{2}$ per cent magnesium, is probably, if not the best, at all events quite as good as any other alloy of that particular type.

The Use of Higher Quenching Temperatures.

The quenching temperatures employed in the course of all the earlier experiments in connexion with alloy " Y," were limited to 480° C. This was done on the basis of the belief that the alloy " Y " was closely analogous in constitution and properties to alloys of the " D " type (resembling Duralumin), and as is well known, alloys of the latter type cannot safely be quenched from a temperature much exceeding 500° C. Simultaneously with the work already described, however, a research was being carried out on the constitution of the alloys of aluminium with nickel and copper, and it was found that when these alloys have attained their stable equilibrium condition, those containing copper up to and over 4 per cent show no critical point corresponding to the melting temperature of the eutectic formed by the compound $CuAl_2$ with aluminium. Even in cast alloys which have been annealed for six hours or more at 480° C., there is no critical point below 554° C. Quenching experiments were also made on small chill cast ingots which had been held at a temperature of 540° for thirty minutes. When quenched in water these showed no sign that any liquid metal had been present at the moment of quenching. Subsequent development of the study of the constitution of the aluminium-nickel-copper alloys has furnished an explanation for these phenomena, since it appears that in the presence of nickel even in smaller proportions than 2 per cent, the compound $CuAl_2$ disappears entirely when equilibrium has been attained. It seems to be replaced by a different body, probably of the nature of a ternary compound of aluminium, copper and nickel, which may or may not have the power of dissolving any excess copper present. In addition, the alloys also contain particles of the compound $NiAl_3$. The microstructure of the " Y " alloy is discussed below. Here it is sufficient to indicate that the

addition of nickel brings about a profound change of structure and constitution in the alloy, which therefore cannot be regarded as analogous to alloys of the " D " type. From the practical point of view this difference of constitution and structure has the important advantage that it is possible with these alloys to employ considerably higher quenching temperatures than those which can safely be used with alloys of the " D " type.

Experiments on the application of higher quenching temperature to alloy " Y " in sheet form, which had been prepared by hot-rolling from slabs 1½ inch thick, were made by holding the material for thirty minutes at 520° C. and quenching it in boiling water. After ageing, the materials so treated gave very uniform results, typical examples of which are given in Table 34.

TABLE 34.

Thickness.	Yield stress.	Ultimate stress.	Elongation.
In h.	Tons per sq. inch.	Tons per sq. inch.	Per cent on 2 in.
0 05	13·4	24·0	20
—	13·3	23·9	20

Comparison of the above Table with Table 31 relating to similar material which had been quenched from 480° C., shows a distinct improvement in ultimate stress and a simultaneous increase in the ductility.

Simultaneously with the work described above, experiments on the use of higher quenching temperatures were also carried out at the Royal Airship Works at Cardington, where work on the manufacture and treatment of alloy " Y " was carried out in close conjunction with that of the present Authors. Experiments on the production of thick wire for the manufacture of rivets out of this alloy at the R.A.W. had indicated the advantage which could be gained from the application of cold-rolling to the treatment of this material. In accordance with this indication, the alloy, after breaking down hot from slabs 1½ inch thick to a thickness of 0·25 inch was

further reduced by cold rolling, applied in stages each entailing a
reduction of approximately 10 per cent in thickness.　At the end of
each stage the alloy was heated to 480° C. and then quenched.　After
being rolled down to the thickness finally required, the material was
subjected to a final heat treatment by heating it to 530° C. followed
by quenching in boiling water.　The resulting sheet has been tested

FIG. 66.—*Alloy " Y."　Effect of varying quenching Temperatures.*
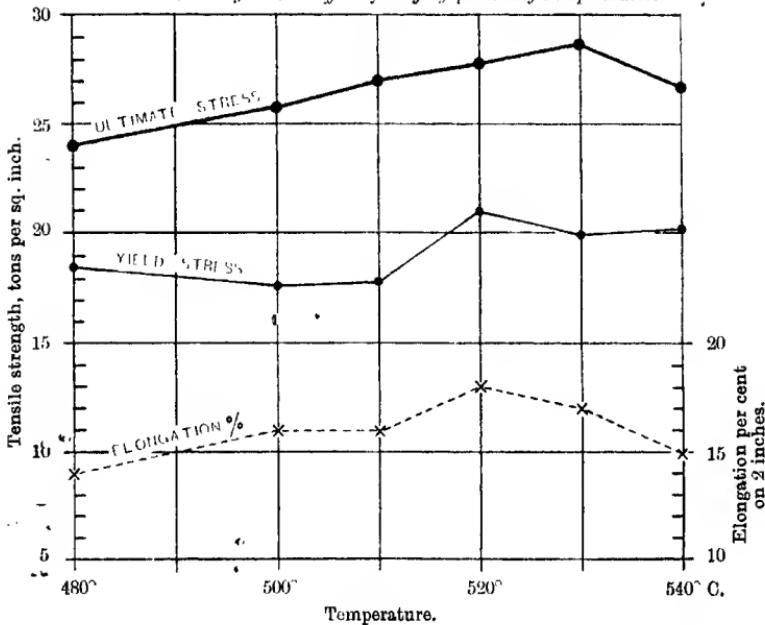
both at the R.A.W. and by the Engineering Department of the
National Physical Laboratory at the Authors' request.　The material
shows an ultimate stress of 26 to 27 tons per square inch with 16
to 18 per cent extension on 2 inches in the case of sheet 0·05 inch
thick ;　thinner sheet 0·18 inch thick shows tensile strength of 27
to 28 tons per square inch with elongation of 15 to 17 per cent on
2 inches.　The effect of varying quenching temperature upon the
ultimate stress, yield stress and elongation of the " Y " alloy prepared

by cold rolling at the R.A.W. is shown in the graph Fig. 66. It
will be seen from that Fig. that the ultimate stress increases
steadily with increasing quenching temperature up to 530° C. but
falls markedly when the quenching temperature is increased to 540° C.
The elongation is not much affected, but a quenching temperature
between 520° C. and 530° C. appears to give the best results.

Microstructure of the Alloys.

The microstructure of an alloy containing four or, if we include
silicon, five essential components is necessarily complex, and its
interpretation taken by itself is practically impossible. The only
satisfactory method of approaching a study of such a case consists
in the systematic investigation of the equilibria and constitution
and structure of the ternary system of alloys principally concerned.
In the present case this is the system aluminium-copper-nickel.
When this system has been thoroughly investigated, at all events
so far as the aluminium-rich alloys are concerned, it should be
possible to ascertain the influence upon selected groups of those
alloys of additions of magnesium or silicon or both. In view of
the very great promise and importance of alloys of the "Y" type this
investigation has been begun in the Metallurgy Laboratory at the
National Physical Laboratory by Dr. Haughton and Miss K.
Bingham. The thermal curves of the aluminium-copper-nickel
system, however, are of a very complicated and intricate
character, and the process of unravelling the difficulties has proved
to be lengthy. The results obtained up to the present time, although
they furnish some indication of the general character of the alloys
in question, are not sufficiently complete to justify their inclusion in
the present Report. Such interpretations, however, as can be
attempted of the microstructure of alloys of the "Y" group, are
necessarily based upon the preliminary results obtained by Dr.
Haughton and Miss Bingham. Some of these have already been
referred to. Study of the ternary system makes it clear that the
addition of even small amounts of nickel to alloys of aluminium
and copper produces a very profound change in their structure and
constitution, resulting, when the alloys are in equilibrium, in the

elimination of the compound $CuAl_2$, and in the presence in the alloys of comparatively large quantities of a ternary body containing both copper, aluminium, and nickel. In addition, the compound $NiAl_3$ also occurs in these alloys.

The microstructure of the alloy " Y," as cast, has already been illustrated in Fig. 23, Plate 3, referred to in Section II (*b*), page 13. The structure shown in that photograph represents crystals of a solid solution consisting mainly of aluminium, together with a network structure of a dark etching constituent which, under a magnification of 150 diameters, merely appears as a dark reticulation. Under higher magnifications, however, it is seen that this network contains two distinct constituents, and possibly more. It seems probable that the bulk of the network is made up of the ternary body referred to above, while the second constituent seen in it may very likely be the compound $NiAl_3$. In what form magnesium may be present in these alloys has not been ascertained so far.

The microstructure of the cast material is very considerably modified after it has been reduced by hot forging to the extent of 10 per cent. The resulting structure is shown in Fig. 67, Plate 9, under a magnification of 45 diameters, and in Fig. 68, Plate 9, under 500 diameters. Further modification of the microstructure occurs when the forged material is annealed at 480° C. The structure resulting from treatment of this kind for thirty minutes followed by quenching in cold water is shown in Fig. 69, Plate 9 (by 150). Here considerable recrystallization has taken place, and in the dark-etching regions the occurrence of two constituents is clearly evident.

" The effect of rolling on the microstructure of the alloy is very marked. This is shown in Fig. 70, Plate 9, under a magnification of 150, taken from a broken down blank 0·75 inch thick, which has subsequently been annealed for eighteen hours at 480° C., with a view to bringing as much as possible of the various compounds into solid solution. In order to retain the compounds in solution, the material has been quenched. It will be seen that in spite of this rather drastic treatment, a large amount of free compound is

still present in the alloy. This is remarkable in view of the comparatively high ductility which material in this condition displays. A sample of rolled alloy when the thickness has further been reduced to 0·2 inch is shown in Fig. 71, Plate 10 (by 150) air-cooled after rolling. The amount of compound which appears to be present in this structure is distinctly larger than that shown in Fig. 68. Plate 9. Finally, in Figs. 72, Plate 10 and 73, Plate 10, both under a magnification of 500 diameters, are shown the structures of alloy " Y " in the finished condition. Fig. 72 relates to material which has been hot-rolled and heat-treated, while Fig. 73 refers to material which has been reduced in the later stages of rolling by cold working and subsequently heat-treated.

For purposes of comparison with alloy " Y," two micrographs are shown in Figs. 74 and 75, Plate 10, which refer to alloys of modified composition described in Table 32. Fig. 75 refers to alloy " W 871 " containing 6 per cent of copper, 3 per cent of nickel, 1 per cent of magnesium, in the condition as cast. Fig. 74 refers to alloy " W 847 " containing 6 per cent of copper, 3 per cent of nickel, and 1½ per cent of magnesium after being hot-rolled down to a thickness of 0·34 inch and annealed for 3 hours at 500° C. It is interesting to note in the latter figure that the elongated crystals formed during rolling have not become equi-axed in consequence of the prolonged annealing to which the material has been subjected. In other respects the alloy " W 847 " is very similar to standard " Y " alloy. In the case of " W 871," Fig.7 5, a strongly marked dendritic structure is apparent, and this is not often seen in castings of alloy " Y " itself.

III(d). *Extrusion.*

In Sections III a, b and c above, an account has been given of the great difficulties which had to be overcome in the early stages of the pro 'uction of both alloys " A " (3/20) and " E." These difficulties mainly centred around the problem of forging and breaking down cast billets and slabs. Once this stage had been overcome, the further rolling gave no great trouble. In searching for a means of overcoming these initial difficulties, it occurred to

the Authors that the extrusion process might provide a means for
bringing these apparently intractable materials either into the
finished condition required for service, or at least into a condition
corresponding to that of the rolled blanks, which would be readily
amenable to further rolling or other treatment. Facilities for
carrying out extrusion experiments were placed at their disposal
with great readiness by the British Aluminium Company, first at
their Milton works, and later at their Warrington works, and also
by Sir Gerard Muntz, Bt., on behalf of the Muntz Metal Company,
of Birmingham. Although, as has been indicated above, the initial
difficulties relating to the forging and rolling of these alloys were
subsequently completely overcome, so that recourse to the extrusion
process cannot now be regarded as essential to the production of
the alloys, yet the large number of extrusion experiments which
have been carried out have yielded results of considerable interest,
and a brief account of them is therefore included in the present
section.

The earlier extrusion experiments, carried out at Milton and at
Birmingham, with a series of alloys of various compositions, showed
that a large number of these alloys could be successfully extruded,
but that for them the temperatures employed in the extrusion of
pure aluminium were much too high, while the pressures required
for the extrusion of these harder alloys were very much greater
than those required for aluminium. As it was anticipated, and
the anticipation was soon verified, that material extruded in
relatively large sections could afterwards be readily rolled to
smaller sections, the largest available dies were then used for the
extrusion experiments. This was done partly also in order
to keep the pressures on the rams in the presses as low as possible.
In order further to diminish the pressures arising in the extrusion
of these alloys, much shorter billets than those ordinarily employed
with pure aluminium were used for the experiments.

A good deal of difficulty was experienced in obtaining extruded
material free from longitudinal defects, leading to a laminated
fracture when a bar was notched and broken by a blow. Further
reference will be made to this matter in connexion with the study

of the microstructure and physical prop. ries of the extruded
material. At an early stage of the extrusion experiments, however,
it was realized that the cause of these defects, which occurred in
many extruded metals beside the special alloys under experiment,
required consic rable investigation. A series of experiments were
therefore carried out in which soft plastic substances built up out
of different coloured layers, were extruded from small model
cylinders. The results of this research will be described elsewhere,
as they have a bearing upon many other matters besides the
extrusion of light alloys. It may be stated here, however, that
while it is not suggested that the data obtainable by the extrusion
of soft plastic material are directly and quantitatively applicable
to the extrusion of much harder materials such as alloys, yet the
results obtained proved a very useful guide in avoiding defects
of the kind indicated.

It appeared perfectly clear from the experiments in question, and
was thoroughly confirmed by subsequent experiments with actual
alloy billets, that the longitudinal central defect leading to lamination
in exc uded rod and bars is due to the drawing down into the bar
of defects originally existing in the ingot, or drawing down into
the bar portions of the skin, particularly from the back of the original
ingot. Two precautions were accordingly adopted which, to a very
large extent, eliminated these defects. In the first place, the ingots
were cast in such a way as to avoid as far as possible the formation
of a contraction cavity. The evil influence of this contraction cavity,
or of the enclosures which are formed when this cavity is filled up
by the ordinary process of " following up " with molten metal, can,
to a large extent, be eliminated by turning the top or feeding end
of the ingot towards the die in the extrusion press. The drawing
down of the rear surface of the ingot into the centre of the bar can also
be avoided by placing between the ram of the press and the billet
or ingot a friction plate. This is a steel plate which on the side
turned towards the ingot is provided with corrugations. These are
driven into the metal of the ingot and prevent the flow of the surface
layers. While these precautions undoubtedly diminish, even if they
do not entirely eliminate, the defects in question, yet in commercially

TABLE 35.

Alloys extruded at Birmingham and Warrington.

Melt No.	Diameter of extruded rod.	Composition.					Behaviour.
		Mg.	Mn.	Si.	Cu.	Zn.	
97-mm. diameter to 1¼-in. diameter (from sand cast 4¼-in. diameter billets).							
Series A. Birmingham.							
W188 . .	1¼ inch	—	—	—	3	25	—
W191 . .	,,	0·25	—	—	3	15	—
W193 . .	,,	0·5	—	—	2·5	13	—
123 mm. to 1½ and 2¼ inch diameter.							
Series B. Birmingham.							
W194 . .	1½ inch	0·5	—	—	.	15	Sound
W214 . . W234 . .	,,	0·5	0·8	—	3	—	,,
W216 . . W217 . .	,,	0·5	0·8	—	4	—	,,
W235 . .	,,	—	—	—	4	15	,,
W236 . .	,,	.	—	—	4	12	,,
W237 . .	,,	0·5	—	—	3	15	
W238 . .	,,	0·5	—	—	3	20	Failed
W239 . .	,,	0·25	—		3	20	
W240 . .	,,	—	—	—	4	20	Sound
W241 . .	,,	—	0·5	—	2·5	20	,,
W242 . .	,,	0·5	0·5	—	2·5	20	Crack at one end
W243 . .	2 inch	—	—	—	3	20	Sound
W244 . .	,,	—	—	. .	3	20	,,
W245 . .	2¼ inch	—	—	. .	3	20	,,
Series C. Birmingham.							
W262 . .	1¼ inch	0·25	—	—	3	15	,,
W263 . .	,,	0·25	—	—	3	15	,,

W471, 473, 474, 476, E, F, G, (H.T.) compositions all failed.

6¾-inch diameter to 4-inch × ¾-inch, and 5-inch × 1¼-inch sections.

Series D & E. Warrington.							
E (H.T.). .	—	0·5	0·5	0·25	2·5	20	—
F (H.T.). .	.	0·5	0·5	0·80	2·5	20	—
G (H.T.). .	—	0·35	0·35	0·80	2·5	18	—

extruded material, where such precautions are not always taken, such longitudinal defects are very frequently found. They are not removed by subsequent rolling or heat treatment of the extruded material, and for that reason the Authors would prefer to see the use of extruded material avoided as far as possible in the case of the light alloys described in the present Reports.

Materials Extruded.

The first series of extrusion experiments at the Milton Works of the British Aluminium Company were confined to billets of alloy " A " (3/20). At a somewhat later date, two lengthy series of billets of varying compositions were extruded at the works of the Muntz Metal Company in Birmingham. Finally, in the latest extrusion experiments, billets of the alloy " E " and its modifications " F " and "G" were extruded at the Warrington works of the British Aluminium Company. The compositions of the billets employed in the Birmingham and Warrington experiments are stated, together with their identification numbers in Table 35. The same identification numbers are used in Table 36, where tensile tests on extruded and on extruded and subsequently rolled material are tabulated. The sizes of billets employed varied necessarily with the dimensions of cylinders and containers available in the various presses. Those employed at Milton measured 5 inches in diameter, while those used in Birmingham measured in one series 97 mm. and in another 123 mm. in diameter. The billets extruded at Warrington, on the other hand, were 6¾ inches in diameter and the experimental ingots were cast to those dimensions. As already mentioned, however, it was necessary to reduce the length of the ingots very considerably owing to the excessively high pressure which would otherwise have been required.

Mechanical Properties.

The results of tensile tests on the material, both as extruded and after further hot-rolling, and in some cases heat-treatment, are given, as already stated, in Table 36.

TABLE 36.—*Tensile Tests.*

Material extruded at Milton, Birmingham and Warrington.

Melt No.	Diameter of Extruded Rod.	Condition.	Yield stress. Tons per sq. in.	Ultimate stress. Tons per sq. in.	Elongation on 2 inches per cent.
		"A" Alloy (3/20).			
E11	1¼ inch	Extruded	15·8	26·1	22·0
E12	2¼ "	"	18·3	27·7	19·0
E12	2¼ "	"	16·6	26·8	20·0
E3	1¼ "	"	15·2	23·8	10·0*
E3	1¼ "	"	15·0	21·3	5·0*
E3	1¼ "	"	12·6	24·3	13·0*
E9	Flat bar 3¼ in. × ¼ in.	"	17·0	27·2	19·0
W181	1¼ inch	"	22·5	30·5	15·0
W183	1¼ "	"	18·6	26·4	20·0
W181	1¼ inch	Annealed 200° C, 30 minutes	17·8 / 18·5	27·4 / 27·2	16·0 / 16·5
HE9	2¼ "	Hot-rolled to 1¼ inch diameter	16·2 / 13·1	25·8 / 27·8	20·0 / 17·0
		Other Copper-Zinc-Aluminium Alloys.			
W189	1¼ "	Extruded	22·9	29·1	16·0
W236	1½ "	Hot rolled to ⅞ inch diameter	7·0	19·0	22·0
W235	1½ "	"	16·7	24·2	17·0
W240	1½ "	"	20·6	27·6	14·0

* Laminated fracture.

TABLE 36—*continued*.

Melt No.	Diameter of Extruded Rod.	Condition.	Yield stress. Tons per sq. in.	Ultimate stress. Tons per sq. in.	Elongation on 2 inches per cent.
		Magnesium-Copper-Zinc-Aluminium Alloys.			
W191	1¼ inch	Extruded	21·3	28·0	17·0
W193	1¼ ,,	,,	17·0	25·6	15·0
W191	1¼ ,,	Do. quenched 450° C. and aged	21·7	31·2	22·0
W193	1¼ ,,	,, ,, ,,	18·5	29·6	22·0
		Magnesium-Zinc-Aluminium Alloys.			
W194	1½ inch	Hot-rolled to ⅞ inch diameter	18·6	24·4	25·0
W194	1½ ,,	Do. quenched 500° C. and aged	13·9	24·4	28·0
		Manganese-Copper-Zinc-Aluminium Alloys.			
W241	1½ inch	Hot-rolled to ⅞ inch diameter	21·3	27·1	16·0
W241	1½ ,,	Do. quenched 350° C. and aged	13·6	27·6	21·0
		Magnesium-Manganese-Copper-Zinc-Aluminium Alloys.			
W242	1½ inch	Hot-rolled to ⅞ inch diameter	23·7	29·7	15·0
W242	1½ ,,	Do. quenched 350° C. and aged	25·7	37·4	13·0
W242	⅞1½ ,,	,, 400° C. ,,	21·6	40·7	9·0

As regards the alloy " A," a tensile strength as high as 30·5 tons per square inch was recorded in one case, while several other tests gave results approaching 30 tons per square inch. In these cases, however, the elongation was always considerably below 20 per cent, suggesting that the material had received a certain amount of cold work during extrusion. This view is confirmed by the microstructure of such material (see below). Broadly speaking, it may be said that the tensile properties of the extruded material, when in a satisfactory condition, are closely analogous to those of the same alloy when obtained by forging and rolling. The occurrence of the undesirable laminated structure, however, is indicated in the Table by a number of abnormally low results.

In addition to alloy " A " (3/20), a number of other alloys of the ternary series, copper-zinc-aluminium, were extruded. These include the alloy " B " (3/25), and also an alloy containing 4 per cent of copper and 20 per cent of zinc. Neither of these, however, appear to present any very considerable advantage over alloy " A " (3/20).

The next series of alloys represent the effect of the addition of magnesium, varying from $\frac{1}{4}$ to $\frac{1}{2}$ per cent, to various members of the ternary copper-zinc-aluminium system. These, even after heat treating and ageing, do not present any material advantage over alloy " A." In the absence of copper (magnesium-zinc-aluminium series) there appears to be no age hardening at all.

The most interesting group of alloys are those of the type of alloy " E." In the extruded and hot rolled condition, after quenching, this alloy in one instance attained the remarkable tensile strength of 40·7 tons per square inch. This is the highest tensile figure recorded in the present Report.

Impact Tests.

A series of notched-bar impact tests have been made on the extruded material obtained from alloy " A " (3/20). For comparison, similar tests have also been made on extruded material of alloy " D," corresponding in composition to " Duralumin." These tests have been made on the Charpy International test-piece, 10 mm. square in section, 40 mm. span, with a notch 5 mm. deep and having

at the root a radius of $\frac{2}{3}$ mm. The energy of the blow was 3·5 kg.n/M. Test-pieces for these tests were cut from the material as extruded in various sizes from $2\frac{1}{8}$ inches to $1\frac{1}{4}$ inch diameter, and both sound and laminated material was employed. The sound material, as will be seen from the Table, gave results of remarkable uniformity, the impact figure being higher for the material extruded to the small diameter. A remarkable difference was, however, observed between the laminated and sound material under this test. In order further to emphasize this difference, test-pieces were cut from the laminated central portion of an extruded bar and from the sound peripheral part of the same bar. It may appear at first sight surprising that the impact figures for the laminated central portions of the bar are in every case considerably higher than for the sound external portion. If, however, the nature of the impact tests is considered, it will be seen that lamination must necessarily give rise to a relatively high impact figure. To make this clear, it is only necessary to consider the case of a test-piece made up of a bundle of thin plates. Such plates, beyond the depth affected by the notch, behave as unnotched specimens and necessarily give an apparently high impact figure.

Extrusion of Alloys "E," "F" and "G."

Some results of extrusion of alloy "E" have already been given above. These refer to round rod, $1\frac{1}{4}$ inch in diameter, which was the only material of this kind successfully extruded. At a later stage, a series of trials were made with the object of extruding flat bars suitable for subsequent rolling into strips as required for the production of structural sections used in aircraft. A first attempt in this direction, made at the Works of the Muntz Metal Company Birmingham, failed completely, showing that the extrusion of these harder alloys was a matter of much greater difficulty than in the case of such materials as alloys "A" (3/20) and "B" (3/25). Later, a very much larger extrusion press, capable of exerting a total pressure of 2,000 tons, became available at the Warrington works of the British Aluminium Company, and further efforts to extrude flat

bars were made with this plant. The die first used gave a bar
4 inches wide by $\frac{3}{4}$ inch thick, but extrusion through this proved
difficult with the harder alloys. Some sound material was obtained,
showing that successful extrusion should be possible, but on the
other hand, the front ends of the bars were in several cases severely
split, and a large amount of tearing and cracking at the edges,
particularly at the corners of the bars, occurred. Somewhat similar,
but still more marked, phenomena of this kind had been met with in
the experimental extrusion of alloy "A." In that case, however,
the violent splitting of the round bar after extrusion was found to

TABLE 37.

0·05 inch thick strip rolled from extruded Flat section 4 in. $\times$ $\frac{3}{4}$ in.

Alloy.	Yield stress.	Ultimate stress.	Elongation per cent on 2 inches.
	Tons per sq. inch.	Tons per sq. inch.	
F	29·2	33·9	15
	28·3	33·6	15
G	24·2	33·1	19
	26·8	32·5	17

be due to the use of an unduly high temperature. It would appear
as if the metal, after passing through the die of the extrusion press,
is under severe lateral compression. If it is too hot, and consequently
too weak to resist this pressure, it is liable to split in a remarkable
manner. It was found in the case of alloy "A" that reducing the
temperature, rapidly diminished this form of failure, until at a
suitably low temperature it was eliminated entirely. In the case
of the harder alloys now under consideration, the temperature used
was 340° C. to 350° C. At lower temperatures the extrusion pressures
were found to be excessively high. The difficulty in this case appears
to arise from friction at the corners of the rectangular section of the
bar, which seems to result in a tearing of the edge at intervals.
As the bar increased in length during the process this tearing

diminished and the later portions of the bars were generally quite sound. Had it been possible to extrude much longer bars, considerable lengths of sound material would have been obtained.

Sufficient sound material of alloys " F " and " G " was obtained to allow of reasonable lengths being rolled into strips at the Cardington Airship Works (now R.A.W.). This strip was afterwards heat treated and tested. The results of tensile tests are given in Table 37.

It will be seen that the tensile tests are slightly lower than those obtained from material of a similar composition prepared by forging and rolling.

TABLE 38.

0·05 inch thick strip rolled from extruded Flat Section 5 in. × 1¼ in.

Alloy.	Yield stress.	Ultimate stress.	Elongation per cent on 2 inches.
	Tons per sq. inch.	Tons per sq. inch.	
E	18·0	39·4	12
	23·1	38·4	11
G	28·0	34·5	17
	21·8	33·3	15

In view of the difficulties experienced in extruding the bars having a section of 4 inches by ¾ inch, it was decided to try extrusion through a larger die. The British Aluminium Company very kindly had a die prepared having an aperture giving a bar of rectangular section 5 inches wide by 1¼ inch thick. With this die it was found that the pressures required were materially lower than with the narrower section, and somewhat more successful results were obtained, although the temperature of extrusion could not be lowered very much. Even with these thicker bars, however, there was still a large amount of cracking at the edges, and particularly at the corners. Again, these defects occurred mainly at the beginning of the bars, and larger amounts of the various alloys were obtained

in the form of sound extruded sections. Examples of such extruded bars having defective ends but subsequently becoming sound, are illustrated in Fig. 76, Plate 11.

The results of tensile tests on 0·05 inch thick strip, rolled in the N.P.L. Mill from this larger extruded section, and heat-treated and aged, are given in Table 38.

It will be seen that the results are somewhat inferior to those obtained from material of similar composition prepared by forging and rolling.

Before the completion of the experiments referred to above, the problem of preparing alloys " E," " F " and " G " in the form of rolled strips from cast and forged material had been solved, and the expensive expedient of installing an extrusion press for the preliminary treatment of these alloys could, therefore, be abandoned. These extrusion experiments were, therefore, not carried beyond the point described above.

Microstructures of Extruded Material.

Microscopic examination of extruded material in the case of all aluminium alloys investigated in the present research, shows at once that the material has undergone a change as profound as that which occurs when a cast slab is rolled and forged. Broadly speaking the microstructure of extruded material is similar to that of metal which has been annealed after hot working. There are, however, a few abnormal features of considerable importance. The typical structure, of alloy " A " (3/20) after extrusion into round rods $1\frac{1}{4}$ inch and $2\frac{1}{8}$ inches in diameter are shown under a magnification of 150 diameters in Figs. 77 and 78, Plate 11 respectively. Annealing at 450° C. produces no appreciable change in such a structure as that shown in Fig. 77. In some cases the crystal boundaries of the material are curiously corrugated as shown in Fig. 79, Plate 12 (× 150). This is taken from a rod $1\frac{1}{4}$ inch in diameter. Fig. 79, and still more Fig. 80, Plate 12, show an abnormal feature which is very frequently seen in extruded rods. This is the presence of large elongated areas enclosed in a well-defined

boundary, and sometimes appearing to be filled with a mass of very minute crystals.

Annealing, even at a temperature as high as 450° C., does not cause these large elongated areas to break up or disappear in any way. Either it produces no effect whatever or, in some cases, it causes the disappearance of the internal network within the elongated " grain." The latter case is illustrated in Fig. 81, Plate 13 (× 150). Fig. 82, Plate 13 is taken from the core of an extruded rod showing a considerable amount of lamination. This flaw appears to contain fragments of metal, but their failure to coalesce on annealing suggests that they are surrounded by layers of non-metallic material such as oxide or dross. The persistence of the elongated " grains " seen in the better samples of extruded rod, may also be due to the fact of their being contained in envelopes of non-metallic matter. They suggest the appearance of " sacks " of crystals enclosed in an envelope which acts as a barrier to recrystallization. These elongated grains are probably milder examples of the defect which in the more serious cases take the form of the flaws illustrated in Fig. 82, Plate 13.

The general features referred to above are common to all or nearly all of the alloys which have been tested under extrusion. In some cases, however, additional features make themselves felt.

In the case of alloy " B " (3/25) a hard dark-etching constituent makes its appearance, and under high magnification this is found to have a finely-laminated eutectoid structure. This is the decomposed β body of the copper-zinc-aluminium system, which here exists in a meta-stable condition. Its appearance is illustrated in Fig. 83, Plate 14 under a magnification of 600 diameters.

The microstructures of alloys of the types " E," " F " and " G," after extrusion and subsequent rolling into strip are illustrated in Figs. 84 and 85, Plate 14 under a magnification of 150 diameters. In the case of Fig. 84, the etching reagent (nitric acid, 25 per cent) was so chosen as to indicate as clearly as possible the presence of the hard dark-etching constituent. In the case of Fig. 85, on the other hand, the etching reagent employed (10 per cent caustic potash) was selected for the purpose of developing the crystal boundaries.

In the latter mode of etching, however, the dark constituent is unduly severely attacked so that it appears to be present in excessive amount.

III(e). *Summaries and Comparative Data for Wrought Alloys.*

In the four preceding sections of the present Report, a large number of data have been given concerning the physical and mechanical properties of the various alloys studied. The necessity of describing these properties for each alloy in turn, however, makes it very difficult to formulate any estimate of the relative values of various alloys when considered from different points of view. In the present section, therefore, the results obtained from each of the alloys in typical tests have been tabulated and summarized as briefly as possible, in order to facilitate comparisons between the various materials.

Tensile Tests.

The properties of alloy " A " as revealed by tensile tests on the material in various sizes and conditions, are tabulated in Table 39. The data there given may be taken as representing mean values derived in many cases from a very large number of tests. Much higher values have been obtained in special instances in nearly every case, but the data given are those which can be readily obtained in ordinary manufacture. For the purpose of drawing up specifications for example, these data might safely be employed, provided that a small margin is left between the specification figures and those given in the Table.

In the above Table, the data referring to " Elastic Limit " state the stress in tons per square inch at which the first departure from Hook's law was observed by means of a delicate extensometer ; the figures, therefore, represent the " Limit of Proportionality." A column has also been included headed " Limiting Proof Stress." This has been done in view of the adoption of the term " Proof Stress" in place of " Yield-Point " for non-ferrous alloys, and other materials not showing the ordinary typical yield-point of mild

TABLE 39.—*Summary Table. Tensile Tests. Alloy 3/'20. Density 3·1.*

Material.	Condition.	Elastic limit.*	Limiting proof stress*	Yield stress.*	Ultimate stress.*	Reduction of area, per cent.	Elongation per cent on 2 in.
Rod. 1½-inch diameter.	(1) Hot rolled.	11·8-13·0		17·6	26·4	29·8	17
,,	,,			18·5	27·4	—	18
,,	,,			18·3	27·0	36·0	21
,,	,,			17·4	26·7	—	30†
1½ ,,	(2) Annealed 200° C.	9·1		15·1	24·8	21·4	21
,,	,, 225°-250° C.				26·1		19
Sheet. 0·104-0·048-inch thick.	(1) Hot rolled.		18-21	16·9	27·2		19
0·020 ,,	,, aged 4 years.				25·2		12-17
0·050 ,,	(2) Do., annealed 200° C.			27·0	33·0		16
0·104 ,,	(3) Cold rolled.			12·3	22·2		24-28
0·119-0·048 ,,	(4) Do., annealed 225-250°C.			26·0	29·7		11
				13·0	24·0		20
Wire.	Produced from ¼-inch rod by cold drawing with intermediate low temperature annealing.						
0·063-inch diameter.	(1) *As received* from Messrs. Wiggins.				27·8		8‡
0·090 ,,					31·0		5‡
0·0125 ,,					27·3		2‡
0·063 ,,	(2) Annealed 250° C.				20·7		14‡
0·020 ,,					22·3		9·5‡
0·0125 ,,					20·7		—

* All in tons per sq. inch.   † On 1 inch.   ‡ On 4 inches.

TABLE 40.—*Summary Table. Tensile Tests. Alloys "E," "F" and "G." Density 3·1.*

Material.	Condition.	Elastic limit. *	Limiting proof stress. *	Yield stress. *	Ultimate stress. *	Reduction of area, per cent.	Elongation per cent on 2 inches.
Rod.							
"E" ⅜-inch diameter.	(1) Hot-rolled.	—	—	23·7	29·7	—	15
,, ,,	(2) Quenched 350° C.; aged 1½ hours.	—	—	16·8	29·7	—	20
,, ,,	Quenched 350° C.; aged 5 days.	22·1	—	32·9	37·8	13·4	12
,, ,,	Quenched 400° C.; aged 5 days.	—	—	21·6	40·7	—	9
"F" ⅜-inch diameter.	Quenched 400° C.; aged.	—	—	25·3	36·0	21·4	16
Sheet.							
"E" 0·04-0·05 inch.	(1) Hot-rolled.	—	—	24·5	32·4	—	18
,, ,,	(2) Quenched 400° C.; aged.	—	28-31	29·3	38·2	—	15
,, ,,	(3) Hot-rolled annealed 250° C.	—	—	14·5	25·3	—	20
"F" ,, ,,	Quenched 400° C.; aged.	—	25-28	30·4	35·4	—	19
"G" ,, ,,	,, ,,	—	24-25	30·6	34·6	—	19

* All in tons per sq. inch.

steel, by the British Engineering Standards Association Committees. The limiting proof-stress as shown, in the Table is the stress which is just insufficient to produce a permanent deformation of $\frac{1}{2}$ per cent on a gauge length of 2 inches. It has been determined by applying successively higher stresses and noting whether or not permanent deformation has been produced at each step. The values given under " Yield Stress " are those obtained by the ordinary method of watching a tensile test-piece carefully with the aid of dividers, and noting the stress at which the first visible deformation takes place. The values are admittedly vague, but are retained for purposes of comparison.

The tensile properties of alloys " E," " F" and " G" are summarized in the same manner as those of alloy " A " in Table 40. Of these, the softest modification, " G," was not available otherwise than in the form of sheet, so that data on rod of this alloy are not included.

It will be seen that these alloys, with varying composition and treatment, give a very wide range of properties, reaching a maximum at 40·7 tons ultimate stress in the case of alloy " E " quenched from 400° C. and aged. It will be seen, however, that this involves the use of a quenching temperature which is rather higher than that regarded as advisable in view of the work on " Season Cracking." It is quite possible, however, that further work will show that quenching from this temperature can be safely employed. Quenching from a lower and perfectly safe temperature such as 350° C. leaves the alloy only slightly weaker (ultimate stress 37·8 tons per square inch), and with a correspondingly greater elongation. There would seem, therefore, to be every advantage in using this lower quenching temperature.

The tensile properties of alloy " Y " are summarized in Table 41.

Finally, in Table 42 typical data for the different alloys in conditions suitable for various uses are placed side by side for comparison.

While a comparison of the kind instituted in that Table is somewhat of an arbitrary nature, since the particular condition selected for each material might reasonably be varied, it yet suffices

TABLE 41.—*Summary Table. Tensile Tests. Alloy " Y " (Density 2·80).*

Material.	Condition.	Elastic limit. *	Limiting proof stress. *	Yield stress. *	Ultimate stress. *	Reduction of area, per cent.	Elongation per cent on 2 ins.
Rod.	Hot-rolled.	—	—	12·0	17·6	30·0	20·0
⅛-inch diameter .	Quenched 580° C. and aged.	7·7	—	15·4	24·1	33·0	23·0
⁷⁄₁₆-inch diameter .	Cold-rolled from 1¼-inch diameter, hot-rolled, quenched 530° C. and aged.	7·7	—	15·2	24·2	34·0	25·0
¼-inch diameter .	Cold-rolled from ½-inch diameter, hot-rolled, quenched 520° C. and aged.	—	—	14·8	22·9	31·0	26·0
Sheet.	Hot-rolled, quenched 480° C.; aged.	—	12·6-14·9	16·2	23·2	—	19·0
0·05 inch	" " 530° C.; "	—	—	17·2	26·0	—	19·0
	Cold-rolled, quenched 530° C.; aged.	9·8	17·5	20·8	26·9	—	17·0
0·018 inch .	" " "	—	19·0	21·0	28·0(1)	—	17·0

* All in tons per square inch.

TABLE 42.—*Comparison of Different Alloys—Tensile Tests.*

Alloy.	Condition.	Elastic limit.*	Limiting proof stress.*	Yield stress.*	Ultimate stress.*	Elongation per cent on 2 inches.	Reduction of area, per cent.
Rod. ⅜-inch diameter							
A (3/20)	Hot-rolled	—	—	18.3	27.0	21	36.0
E (H.T.)	,,	—	—	23.7	29.7	15	—
F (H.T.)	,,	—	—	—	—	—	—
Y	,,	—	—	12.0	17.6	20	30.0
E (H.T.)	Hot-rolled, quenched and aged	22.1	—	32.9	37.8	12	18.4
F (H.T.)	,, ,,	—	—	25.8	36.0	16	..4
Y	,, ,,	—	—	15.4	24.1	23	33.0
D	,, ,,	—	—	14.0	25.1	30	35.0
Sheet. 0.05 inch thick							
A (3/20)	Hot-rolled	—	18-21	16.9	27.2	9	—
E (H.T.)	Hot-rolled, quenched and aged	—	28-31	29.3	38.2	15	—
F (H.T.)	,, ,,	—	25-28	30.4	35.4	19	—
G (H.T.)	,, ,,	—	24-25	30.6	34.6	19	—
Y	Hot-rolled, with cold rolling in final stages, quenched and aged	—	17.5	20.8	26.9	17	—
D	,,	—	—	14.3	27.1	22	—

* All in tons per square inch.

to show the typical differences between the various alloys described in the present Report, and their relative position as compared with the well-known alloys of the " Duralumin " type. This comparison is so easily made in the Table that no further comment is required. From the point of view of tensile strength alone, however, it is possible to express the relative values of the different alloys, in terms of what has been defined in the Tenth Report to the Alloys Research Committee, as " Specific Tenacity," which is the tensile strength (ultimate stress) in tons per square inch divided by the weight of one cubic inch of the metal in pounds. Expressed in another way,

<div align="center">TABLE 43.</div>

Alloy.	Condition.	Specific tenacity.	Miles supported vertically.
A 3/20 . .	Rolled, 29 tons per sq. inch.	262	9
E (H.T.) .	Rolled, heat treated, 39 tons per sq. inch.	360	12·6
Y . . .	Rolled, heat-treated, 27 tons per sq. inch.	268	9
Duralumin .	Rolled, heat-treated, 27 tons per sq. inch.	268	9
Steel . .	Rolled, 30 tons per sq. inch.	105	3·7

this same property can be stated in terms of the length of a rod or wire of the material which, when hanging vertically, is just able to support its own weight. These data are given in Table 43.

Strength at Low Temperatures.

In view of the low temperature to which aircraft structures, etc., are sometimes exposed in service, tests have been made on the strength of various wrought alloys described in the present section of this Report, at low temperatures obtained by immersing them in frozen carbon-dioxide. The tests were made after ten minutes exposure to this low temperature ($-80°$ C.). The results

are shewn in Table 44, which shows the results of tensile tests
at the ordinary temperature and at $-80°$ C. placed side by side for
comparison. It will be seen that alloy " A " is slightly stronger and
harder at the low temperature, than av the ordinary temperature.
In the case of alloy " E " there is only a very slight increase in
tensile strength, but the elongation is slightly larger than at the
ordinary temperature. In the case of alloy " D," corresponding in

TABLE 44.

Alloy.	Test temp.	Yield stress.	Ultimate stress.	Elongation per cent on 2 in.
		Tons per sq. in.	Tons per sq. in.	
A (3·20)	15° C.	16·70	26·60	18·0
			30·00	13·5
	−80° C.	18·90	29·30	13·0
F	15° C.	21·55	40·65	9·0
		33·10	42·90	12·0
	−80° C.	35·40	43·80	12·0
D	15° C.	14·20	25·30	25·5
	−80° C.	13·65	26·10	26·5

composition to Duralumin, there is practically no difference. In
no case can it be said that the alloys are appreciably affected
by the low temperature.

Fatigue Ranges.

The fatigue ranges of a whole series of alloys have been
determined by the Engineering Department of the Laboratory by
means of the Wöhler test, using a tubular test-piece. The
determinations have been made both at the ordinary temperature
(20° C.), and also at a slightly elevated temperature (150° C.), in
view of the possible use of these alloys for the working parts of
machinery exposed to slightly elevated temperatures, such, for
instance, as the connecting rods of aeronautical or automobile

engines. The results of the tests, expressed in terms of the safe
range of alternating stresses which the material might, as the result
of the Wöhler test, be expected to endure indefinitely, are stated
in Table 45. The safe ranges are expressed in tons per square
inch, representing the limits of alternating compression and tension.

The alloys included in the above Table are not only those which

TABLE 45.

Fatigue Tests (Wöhler Method) on Wrought Alloys.

Material.	Safe range of alternating stresses.	
	20° C.	150° C.
Hot-rolled Rod, $\frac{3}{4}$ inch diameter.		
Alloy A (3/20)	$\pm\ 8\cdot7$	$\pm4\cdot5$
Alloy B (3/25)	$\pm10\cdot5$	--
Alloy E*	$\pm\ 9\cdot7$	$\pm5\cdot1$
Alloy Y*	$\pm10\cdot2$	$\pm8\cdot4$
Alloy Y_1*	$\pm10\cdot3$	$\pm8\cdot2$
Alloy Y_2*	$\pm\ 9\cdot0$	$\pm6\cdot5$
Alloy D_1*	$\pm10\cdot5$	$\pm7\cdot2$
Alloy D_2*	$\pm10\cdot5$	$\pm7\cdot9$
Alloy D_3*	$\pm10\cdot2$	$\pm5\cdot0$
Alloy M	$\pm\ 9\cdot0$	±7.0
Alloy R	$\pm\ 6\cdot5$	$\pm5\cdot5$

* Heat-treated and aged.

have already been described in the earlier sections of the present
Report, but a number of others have been included for comparison,
or for the purpose of studying the effects of minor modifications
of composition, which, it was hoped, might favourably influence
their resistance to fatigue, particularly at elevated temperature. The
compositions of the alloys " A," " B," " E," and " Y" are as stated
in the earlier portions of the present Report. Alloy " Y_1" corresponds
in composition with the normal alloy " Y," but with the addition

of $\frac{1}{2}$ per cent of manganese. Alloy "Y_2" has the same addition of manganese as alloy "Y_1," but also an addition of 0·75 per cent of silicon. It will be seen that neither of these additions have improved the fatigue resistance of the material at high temperatures. For purposes of comparison, three alloys of the "D" type, resembling Duralumin, have been included. The compositions of these alloys are similar to those of Duralumin, but in the case of alloy "D_1" the proportion of copper present was 3 per cent, in "D_2," 5 per cent, and in "D_3," 6 per cent It was hoped that possibly the increase of copper content would stiffen the alloys at high temperature as against fatigue stresses, much in the same way as an increase in copper content appears to stiffen castings under the tensile test at high temperatures. It will be seen, however, that the actual results have been the reverse, at any rate, when the alloy with 6 per cent of copper is compared with the softer alloy containing only 5 per cent of copper. Alloy "M" contains copper 2 per cent, nickel 1·5 per cent, magnesium 1 per cent, with the usual impurities (iron and silicon). This corresponds very closely to the alloy sometimes known under the name of "Magnalite." Alloy "R" contains 6·15 per cent of magnesium and the normal impurities (iron and silicon). It will be seen that neither of these alloys afford any exceptionally useful properties in regard to fatigue resistance.

A general comparison of the figures given in the above Table 45 serves to establish the fact that from the point of view of fatigue resistance, particularly at somewhat elevated temperatures, the alloy "Y" is superior to any of the others which have been tested. It may perhaps be suggested that the difference between a material having a fatigue range at 150° C. of $\pm$ 7·9 tons per square inch and one having a fatigue range of $\pm$ 8·4 tons per square inch is not very great. It must be pointed out, however, that the results of the present research as embodied in the above Table make it very clear that even small differences in the fatigue range at elevated temperature are difficult to obtain ; even comparatively large changes in composition appear to affect this property to a very limited extent, so that the difference in question, amounting to about 6 per cent in favour of alloy "Y," as

against the best of the other alloys shown in the above Table, is
of some value. Another result shown in the above Table, although
not of a particularly satisfactory nature, is the fact that the very much
stronger alloys of type "E," which under the ordinary tensile test
show so great a superiority over such materials as "A," "Y" and "D,"
give disappointing results as regards resistance to fatigue stresses,
even at the ordinary temperature. The real reason for this remarkable
discrepancy in properties has not been established, and is a matter

TABLE 46.

Compression Tests on Channels.

Alloy.	Load.	Buckling stress.
	Tons.	Tons per sq. inch.
Alloy Y . . .	1·171 1·259	13·20 14·16
Duralumin (Vickers) .	0·992 1·161	11·15 13·07
Alloy A (3/20) . .	1·469 1·601	.16·50 18·02
Alloy G . . .	1·629 1·719	18·31 19·32
Alloy F . . .	1·862 2·130	20·97 23·90
Alloy E . . .	2·020 2·100	22·72 23·62

calling for further investigation. In this connexion it may be
desirable to point out that the behaviour of these alloys under
direct compression is quite in accordance with their behaviour
under tension, so that weakness in compression alone cannot be
cited as the cause of failure under fatigue.

Compression Tests.

Results of compression tests made on specimens prepared from
hot-rolled rod of alloy "A" have already been described. Actually,

however, light alloys, as a rule, are exposed to compression stresses mainly in the form of channels or other sections such as are employed in the construction o spars or airship girders. When, therefore, material in the form of channels and oth r sections became available for experiment, compression tests on the material in these forms were undertaken. The tests were carried out at the Royal Airship Works at Cardington, and the results are shown in Table 46. The material used was channel, which had been produced from rolled strip by bending in a roll-bending machine. It had a thickness of 0·018 inch. Of this channel, 6-inch lengths were taken and the ends squared off parallel to each other and at right angles to the centre line of the channel. They were tested with free ends between ball centres placed at the centres of gravity of the channel sections ;

Fig. 86.
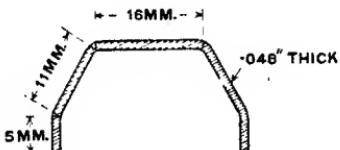
the lengths between the ball centres was $7\frac{1}{4}$ inches, and the cross sectional area of the channel 0·089 square inch. A section of the channel is shown in the sketch, Fig. 86.

It will be seen that the results shown in the above Table bring out in a remarkable manner the superiority under this kind of compression stress of the harder and stronger alloys whose development has been described in the present Report. For purposes of comparison, a sample of Duralumin (manufactured by Messrs. Vickers, Ltd.) has been included in the test. It will be seen that the buckling stress of the channel of alloy " E " is approximately double that of this comparison specimen. Even alloy " Y," whose tensile strength is not very much higher than that of the sample of Duralumin in question, shows a distinct advantage under compression. Since it is generally admitted that in the construction of rigid airship structures particularly, failure usually occurs by

buckling under a compression stress of the kind employed in these tests, the importance of these results is not to be underrated.

General Properties.

Apart from the quantitative data set out for purposes of comparison above, a large number of properties of these various alloys have been described and discussed throughout the present section of the Report. On the basis of these general qualities, it is not easy to form a definite comparative opinion. There can be no doubt that as regards ease of manufacture alloys " A " and " Y " are distinctly superior to those of the type " E," " F " and " G." On the other hand, the latter alloys possess certain valuable properties which may, perhaps, make it worth while to overcome the slightly greater manufacturing difficulties connected with their production. That all the materials in question can be, and actually have been, produced in industrial practice, establishes the possibility of employing them for engineering purposes where their properties are particularly useful. The limitations as to careful treatment which are required by all those alloys needing heat treatment, and also by alloy " A " in order to safeguard it against the risk of " season cracking," have been fully discussed above. Probably the most serious limitation which is likely to arise in connexion with alloys " A," " E," " F " and " G," is that relating to their tendency to undergo serious corrosion if exposed to sea water or other severe conditions. The necessity of protecting them carefully from such exposure, either by selection of the circumstances in which they are employed, or by covering them with a satisfactory protective coating, has already been indicated. It must be borne in mind that even steel is liable to serious and rapid corrosion under severe conditions, yet this limitation has not seriously hindered the application of steel in an enormously wide range of engineering uses. On the other hand, there can be no doubt that alloy " Y," while still possessing valuable mechanical properties comparable, if not superior to, the best of those hitherto obtained with other aluminium alloys, combines with them a very remarkable resistance to corrosion. Another consideration,

which may perhaps govern the choice of alloys for various purposes, is that of their actual density. From the point of view of relative strength and weight of the material, Table 13 above, relating to the specific tenacities of the alloys, gives data upon which a rational choice can be made. At the same time, it must be recognized that for certain purposes a lighter alloy, even if specifically weaker, may be preferable in view of the fact that it can be employed in thicker sections which are not so liable to local damage, etc. From this point of view, the slightly higher density of alloys " A," " E," " F " and " G" is a correspondingly slight disadvantage ; on the other hand, alloy " Y " is as light as any of the stronger aluminium alloys which have yet been produced.

Section IV.

Permanence of the Alloys.

(a). Ageing.

A certain amount of doubt has been thrown, from time to time, upon the reliability and permanence of aluminium alloys. The very extensive and successful use of certain alloys for aircraft and other purposes connected with warfare has, during the past few years, to a large extent removed the doubt. At the same time the phenomena, which have now become well known, of age hardening which occur in certain aluminium alloys, notably those containing magnesium, suggest that these alloys are capable of undergoing changes at the ordinary temperature which bring about very important modifications in their physical properties. Whatever interpretation may be put upon these phenomena, they undoubtedly indicate that most aluminium alloys at ordinary temperatures exist in a condition in which internal changes and rearrangements can and do occur. The question is, therefore, a natural one whether in the course of time these changes, which at first produce results markedly beneficial to the physical properties of the material, may continue or may alter in character with the effect of bringing about a deterioration. Throughout the present research the

Authors have kept these questions before them; and a number
of special experiments have been made with a view to testing what
changes, if any, are undergone by some of the more important
aluminium alloys in the course of time. The very prolonged period
of time through which the present research has been extended, has
enabled them to make tests upon the same material at intervals in
some cases of more than ten years. The evidence which they are
able to present as to the permanence and stability of the alloys is
therefore particularly important. It will be seen, on perusing the
present Section, that on the whole the evidence indicates very
clearly that the alloys, while undergoing certain changes which
generally tend in the direction of increasing the hardness of the
material, do not deteriorate to any appreciable extent within the
period of time covered by the observations, while in regard to
retention of shape and dimensions they are adequately stable for
all practical purposes, with the possible exception of such uses as
accurate standards of length.

The changes which are found to occur in certain of those alloys
which have been studied for a sufficient length of time, all appear
to lie in the direction of an increase of hardness, together with a
slight diminution in ductility as shown by extension under a tensile
test. This gradual hardening in the course of time, appears to be
somewhat analogous in nature to the age hardening which occurs
usually in four or five days in certain alloys after they have been
quenched from a suitable temperature. The nature and mechanism
of this age hardening has been very fully studied and discussed by
the Authors, an account of their work being given in Section V d
of the present Report (see page 235). As is there shown, it appears
to be due to the gradual separation from a state of solid solution
which is stable at a high temperature, of a dissolved substance in
a very finely dispersed condition. The condition that such a change
can occur in an alloy after more or less rapid cooling, is that the
dissolved substance in question shall be more soluble at a high
temperature than it is at the ordinary temperature. When this is
the case, relatively rapid cooling will retain more of the dissolved
substance in solution than can remain in that condition in a stable

state, and there is a tendency for some of it to become separated in a finely divided condition, thus producing gradual hardening of the alloy. It appears probable, although it has not yet been definitely established, that the relatively slight amount of ageing which occurs in such material as alloy "A" in the wrought condition, and in somewhat similar alloys in the cast condition, is due to the separation of the copper-aluminium compound $CuAl_2$. If this is correct, the amount of age hardening which will take place in a given specimen of alloy depends very largely upon the rate of cooling to which it was exposed after last being heated to a temperature in the neighbourhood of or above $400°$ C. It is not surprising, therefore, to find that various samples of the same alloy give somewhat different results in regard to ageing, since the exact conditions of cooling have not been regulated specially from this point of view.

(1) *Cast Alloys.*

Attention was first drawn to the fact that the physical properties, as revealed by tensile tests, of certain cast alloys undergo gradual change in course of time, when it was sought to establish the physical properties which could reasonably be demanded in a specification for a casting alloy for general purposes. Reference to these physical properties has already been made (see page 16). It so happened, in the earlier stages of the present research, that test-piece castings had been prepared at a time when it was not possible to carry out tensile tests upon them immediately, with the result that in some cases tensile data had been recorded from samples which had been stored for several months before being tested. When it was subsequently attempted to repeat these results by means of test-pieces cut from freshly cast material, it was found impossible to do so, and on looking into the matter it was quickly discovered that the alloys undergo a decided improvement in tensile strength as the result of a few months' ageing. The alloys in question are those containing from 2 to 3 per cent of copper and 12 to 15 per cent of zinc, including the alloy now generally known as "L5," and another (alloy "C") containing 3 per cent of copper and 15 per cent of zinc.

The results of tensile tests on sand castings of these two alloys when freshly made and after seven to ten months' ageing are tabulated in Table 47.

Inspection of the above Table shows that in five out of the six

TABLE 47.

Age.	Yield stress. Tons per square inch.	Maximum stress. Tons per square inch.	Elongation per cent or 2 inches.
Alloy "L5." Cast 1 in. diameter.			
9 days	6·8	8·5	3·5
9 days	7·9	9.7	2·5
10 months	11·4	12·1	—
Cast to shape.			
9 days	8·7	11·8	2·0
9 days	8·7	11·8	2·0
10 months	12·3	12·7	1·0
Cast 1 in. diameter.			
9 days	6·0	9·3	2·0
7 months	11·8	12·4	1·0
Alloy C (3/15.) Cast 1 in. diameter.			
½ day	4·5	10·8	2·5
8 months	8·8	13·7	—
Cast vertically ⅝ in. diameter.			
6 days	7·1	13 2	2·5
9 months	15·8	17·0	2·0
Cast vertically 1 in. diameter.			
6 days	7·4	13·2	3·0
9 months	11·3	17·4	3·0

cases ageing has resulted in an improvement in tensile strength of the order of 80 per cent. Little can be said about the change, if any, which may occur in ductility, as the amount of extension in these samples is so small that it cannot be relied upon to furnish

data sufficient y accurate for compa ison. It is interesting to note that in one case the ir provement in tensile strength only amounted to 8 per cent, but in tl t instance the tensile strength of the freshly cast material was appreciably higher than in the others. In this particular case the test-pieces are described as having been cast to shape. This means that instead of being cast in the shape

FIG. 87.--*Hardness (Brinell), Ageing.*
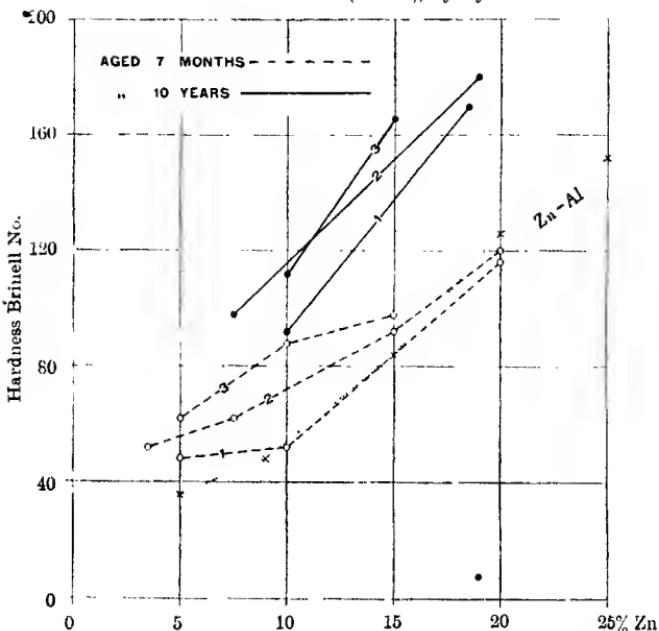
of parallel bars, from which test-pieces with screwed ends were afterwards machined, these particular samples were cast in the shape of shouldered test-pieces which only required slight machining for purposes of accurate finishing. The higher tensile strength of the freshly cast material thus cast in a thinner section is, of course, in accordance with general experience with these a .d indeed with most alloys. The fact that subsequent ageing does not improve these results to the same extent as those obtained with thicker

castings, suggests that the rate of cooling of the thinner pieces has been such as to bring the material already into a condition approximating to that which it would otherwise attain as the result of ageing. Much more complete investigation of the whole phenomena of ageing, however, is required before it is possible to offer any more detailed interpretation of this example.

(ii) *Wrought Alloys.*

Reference has already been made in Section III (*a*) of this Report, dealing with the preliminary investigation of ternary alloys of copper, zinc and aluminium, to changes in the elastic limit, ultimate stress and elongation, which certain of these alloys in the hot-rolled condition have undergone as the result of ten years' ageing. The changes there shown indicate a distinct improvement in tensile properties accompanied by only a comparatively small reduction of elongation. Similar series of changes are shown by hardness tests made by the Brinell Ball method, results of which are incorporated in the graph of Fig. 87, (page 159). In that Figure the Brinell hardness, both of the alloys when aged seven months and when aged ten years, is plotted against zinc content. It will be seen that all the alloys undergo very definite amounts of hardening, and that on the whole the amount of this hardening is greater with high zinc content. These figures, however, relate to the difference between the alloys after seven months, and ten years' ageing, and considerably greater differences would probably be found if the comparison were between material aged ten years and that freshly made. The above hardness tests refer in all cases to material which has been hot rolled to $1\frac{1}{4}$ inch diameter. It will be seen below that material in this condition undergoes ageing to a considerably less marked extent than the same alloys in the form of thin sheet.

The whole question of ageing, particularly of hot-rolled material, has received much more attention in the case of alloy "A" (3/20) where tests, both of freshly made material and of material aged for various periods up to 1,750 days (nearly five years) are available. These tests, which are tabulated in Table 49 relate to the

TABLE 49.—Alloy "A" (3/20)—Ageing Tests. Tensile Tests.

	Thickness or diameter. Inch.	Months aged.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation per cent on 2 inches	Months aged.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation per cent on 2 inches	Months aged.	Yield stress. Tons per square inch.	Ultimate stress. Tons per square inch.	Elongation per cent on 2 inches
							Thin sheet. 0·052-0·054.						
303AL*	0·052	1½	21·3	29·0	16	5¼	19·9	29·8	17	49½	25·1	32·5	1?
8*	—	,,	16·4	29·0	13	5¼	22·6	29·5	13	49¼	24·7	31·4	a?
306B	0·052	,,	18·9	27·8	18	—	—	—	—	49¼	23·6L	33·4	15
310AL* T	0·054	nil	17·3	23·8	20	2¼	19·7	25·8	17	49¼	23·4T	32·2	17
		,,	16·7	26·8	15	2½	18·2	27·7	19		—	—	—
							Thick sheet. 0·103-0·115.						
246L	0·103	nil	19·4	25·7	14	—	—	—	—	56¼	23·2	28·7	16
										56½	23·3	29·1	8
271T	0·114	¾	16·0	26·6	24	9¼	18·1	27·3	20	—	—	—	—
328L	0·103	nil	15·2	29·4	18	2¼	20·1	28·7	18	—	—	—	—
T	0·103	,,	17·5	28·8	18	2¾	19·8	28·4	19	—	—	—	—
							Rod. 1¼-inch diameter.						
176	—	nil	17·0	27·0	20	14½	17·3	26·4	18	58½	22·1	27·9	13
178	—	,,	15·0	25·8	19	—	—	—	—	—	—	—	—
182	—	,,	14·0	25·2	19	14½	19·5	27·2	17	52½	22·3	28·4	14
185	—	,,	16·8	26·0	20	14½	19·4	27·6	16	—	—	—	—
							Bar. 1¼-inch × 1½-inch.						
851	—	nil	14·9	25·8	22	4	15·7	26·2	20	47½	20·7	27·4	17
863	—	,,	15·3	26·4	21	4	17·6	25·8	21	—	—	—	—
370	—	,,	16·0	26·7	19	4	16·0	25·6	18	47½	20·2	26·5	17

* T = transverse. L = longitudinal.

alloy in the form of very thin sheet (0·05 inch thick), thicker sheet (0·01 inch thick), rolled rod 1⅝ inch diameter, and rectangular bars 1¼ inch wide by 1⅛ inch thick.

Inspection of the Table shows at once that ageing is much more marked in thin sheet than in the thicker material. It is quite possible that this difference in behaviour on ageing is connected with the fact that thin sheet when hot rolled necessarily undergoes much more rapid cooling than thicker sheet or than relatively thick bars. The actual ageing effect in the thinnest sheet, thickness 0·05 inch, is very marked. The material as rolled varies in tensile strength from about 27 to 29 tons per square inch, with elongations varying from 15 to 20 per cent on 2 inches. After several years' ageing, the average ultimate strength of the thin sheet rises to well above 32 tons per square inch, together with a still more marked rise in the yield stress. Although, as has been pointed out in the introduction to the present Report, these values for yield stress cannot be regarded as having any very exact significance, the rise in the figures shown is so marked that it clearly indicates a hardening of the alloy and a considerable increase in its elastic range. This very marked increase in tensile strength and stiffness is accompanied by a relatively small reduction in the ductility as indicated by the elongation on 2 inches. The lowest value found in the aged material of thin sheet is 13 per cent, the average being 15 per cent, as compared with a minimum of 15 per cent and an average of just over 17 per cent in the relatively fresh material. These changes in tensile strength and ductility are comparable with those undergone in a few days or even hours by alloys containing magnesium which have been suitably quenched.

In the thicker sheet, thickness 0·1 inch, the results available are not so strictly comparable because the tests on aged material are not made on actual pieces of the same sheet as those representing the condition immediately after rolling, except in the case of one sample which unfortunately had an abnormally low ductility (elongation) when freshly made. In this latter sample the elongation in the aged material fell as low as 8 per cent, but in view of the circumstance just named, this result may be regarded as abnormal.

In the case of circular rod and rectangular bar of greater thickness there is still an appreciable hardening evidenced by increase in ultimate strength and yield point, the former rising from an average value of 26 tons per square inch to the neighbourhood of 27·5 tons per square inch. There is a corresponding but not very large fall in the ductility. Comparison of the figures obtained from material which has been aged for a period in the neighbourhood of one year with the results obtained from ageing of nearly five years, suggests that the ageing process is a very slow one and that it certainly has not been completed at the end of the first year. It appears probable, however, that at the end for five years the ageing process, if not entirely completed, must at all events have slowed down so much that further changes can scarcely be of any considerable importance

(iii). *Other Alloys.*

Alloys " E," " F " and " G," and Alloy " Y " represent such comparatively recent developments in the work of the present Authors, that data concerning their behaviour in regard to ageing are not yet available. Samples of early material of these compositions are, it is true, available which are considerably more than twelve months old. Material produced in the early stages of development of a new type of alloy, however, cannot be regarded as satisfactorily representing the properties and behaviour of that material, and for that reason ageing results on samples of this kind are not included in the present report. So far as alloys " F " and " G " are concerned, it is probable, from their similarity in constitution to alloy " A," that a certain amount of gradual hardening is likely to occur. In view of the fact, however, that these alloys are already age hardened after quenching a few days after their production, it does not appear likely that the increase in hardness due to prolonged ageing will be very appreciable. The same considerations apply to a large extent to alloy " Y."

From the point of view of the permanence and ageing properties of alloys which have been age-hardened after quenching, some data are available regarding tests on pieces of Duralumin cut from

the same piece of channel, after an interval of nearly ten years.
The material in question forms part of an early attempt at the
construction of rigid airships in this country, and was manufactured
by the regular makers of Duralumin.

The results of the tests in question are shown in Table 50.

TABLE 50.

Thickness of material.	Yield stress.	Ultimate stress.	Elongation per cent on 3 inches.	Date of test.
	Tons per sq. in.	Tons per sq. in.		
0·071 inch.	16·9	26·5	16·6	October 1911.
0·070 inch.	17·5	27·5	20·3	
0·070 inch.	20·6	28·7	15·3	January 1921.
0·070 inch.	20·7	27·7	14·7	

It will be seen from the above Table, that although the tests
of the material varied slightly in duplicate samples, there is a definite
and appreciable change in the tensile properties of the alloy after
a period of ageing close upon ten years. It may be added
that the material in this case was stored in a cupboard in the
Laboratory and has been entirely unaffected by corrosion, vibration
or other outside influences. It will be seen that there has been a
marked increase in yield stress. Although no very great amount
of weight can be attached to these figures, yet particularly in
the case of Duralumin, this determination is sufficiently accurate
to show that the differences between the tests of 1911 and 1921,
as stated in the above Table, are real and indicate a stiffening of the
material. This is further borne out by a slight but quite definite
increase in ultimate stress, and a small but equally distinct decrease
in the elongation. It must be pointed out that these changes do
not in the least suggest any serious deterioration in the properties
of the alloy ; in fact, from some points of view, the aged material
is better than that freshly made. On the other hand, these results

serve to show quite clearly that in Duralumin, as in other alloys described in the present Repon , internal changes may and do occur over a period of many years after manufacture, and it has yet to be ascertained whether in any of these materials these changes finally come to an end, or whether they continue indefinitely but slowly and in the same direction as that indicated by these tests, or whether, as is quite possible, the direction of the change is ultimately reversed and the alloys then tend to become softer instead of harder. Fortunately, from the practical point of view, it is evident that these changes are very slow indeed, at all events after the first few years, so that except for structures in which permanence for a large number of years is essential, there need be no hesitation in relying upon the physical constants of these materials as determined a short time after their manufacture.

IV(b). *Stability of Dimensions.*

At the time of the publication of the Tenth Report to the Alloys Research Committee, where the remarkable properties obtainable in castings of zinc-aluminium and copper-zinc-aluminium alloys were pointed out, some doubt appeared to exist as to the permanent stability of such castings. Certain die-castings, described as " aluminium castings," had been submitted to the Authors for examination in a condition which made it clear that the material had undergone serious changes of dimension, leading ultimately to complete disintegration. Although it was found that these castings consisted mainly of zinc, the question of the permanence of aluminium-zinc and copper-aluminium-zinc alloys in the form of castings particularly, appeared to require investigation, if only in order to dispel any doubt which might interfere with the extended practical utilization of these materials. The behaviour of the alloys rich in zinc has also been studied, but the results of that investigation are described elsewhere.*

Early in 1913, arrangements were made for the preparation of

* " Zinc Alloys with Aluminium and Copper." Rosenhain, Haughton and Bingham. Journ. Inst. Metals. Vol. xxiii, No. 1, 1920.

a series of castings which could be used for observation in regard
to possible dimensional changes over a period of years.   By the kind
co-operation of Mr. William Mills, of the Atlas Aluminium Works,
Birmingham, the production of these castings was very much
facilitated, as they were prepared in the foundry of that firm under

FIG. 88.
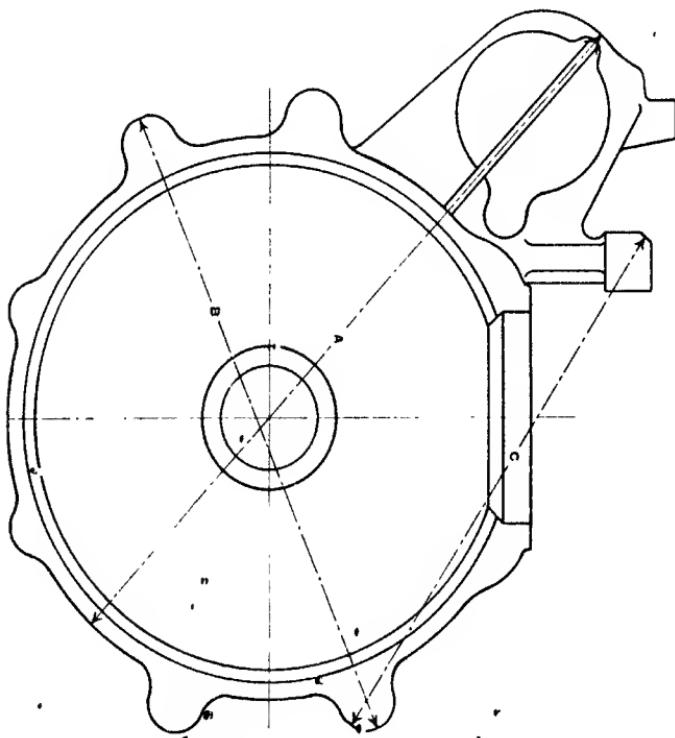
the personal supervision of one of the present Authors.   In order
to test the effect of possible minute changes in the alloys leading
to slight warping in castings, it was decided to use a moderately
complicated casting embodying portions of varying thickness
and of such a shape that warping should produce an appreciable
effect.   The pattern ultimately selected was a motor cycle crank

case of the ordinary type. A plan view of this crank-case is shown
in Fig. 88. For purposes of measurement of dimensions, the
three lines indicated by the letters A, B and C in the Figure were
selected as likely to afford the most severe test of any minute
dimensional changes. In order to allow of accurate measurement,
small flats were machined in the positions corresponding to the ends
of these lines, thus allowing end measurements by means of screw
micrometers to be made. These flats were machined with the
greatest possible care in the Engineering Department of the
Laboratory, but it was found on measurement that it was necessary
in order to obtain consistent values with the delicate micrometer
employed, to secure that the contact pieces of the micrometer
touched the flats always at the same point. In order to secure
this result, small circles were marked on the machined flats and
measurements made within these circles.

In order to ascertain whether exposure to the weather as
compared with mere storage under cover would affect the behaviour
of the castings, two complete series ware prepared. The castings
of one series were exposed on the roof of the Metallurgy Department
of the Laboratory, while the other series were kept in a store room.
In both cases the flats machined for measurement purposes were
kept covered with a coating of vaseline. This was found to afford
complete protection from corrosion and other visible damage to the
flats and could easily be removed when measurements were to be
made.

The material used included a range of alloys containing 20 per
cent of zinc both with and without a number of intentionally added
impurities. A second series included alloys containing in addition
to 20 per cent of zinc, 3 per cent of copper and a similar variety of
impurities. For further purposes of comparison, an alloy containing
4 per cent of copper only was employed. Finally, in order to
ascertain whether dimensional changes, if any, were to be traced
to the alloys or to the added impurities, or whether they must
be attributed to the behaviour of aluminium itself, four
castings were made of aluminium alone. Three of these were
made of the purest available aluminium (99·6 per cent), which for

TABLE 51.

Series.	No. of Case	Percentage Composition.									
		Al.		Zn.		Cu.	1 b.	Fe.	Si.	Mn.	Mg
		Pure.	Coml	Pure.	Coml.						
I. Al-Zn (20 per cent Zn)	1	80·0	—	20							
	2	79·0	·	20	—	· ·	1				
	3	79·0	—	20	—	—	—	1·01			
	4	79·0	—	20	—	—	—	—	0·99		
	5	79·0	—	20	· ·	—	—	—		1·00	
	6	79·5	—	20	—	· ·	—	—	—	—	0·46
	7*	80·0	—	20							
	8†	80·0	—	20							
	9	80·0	—	· ·	20						
	10	—	80	20							
II. A-Zn-Cu (3/20)	12	77·0	—	20	· ·	3·0					
	13	76·0	·	20	—	3·1	1				
	14	70·0	—	20	—	3·0	—	0·84			
	15	76·0	—	20	—	3·0	—	—	0·91		
	16	76·0	—	20	—	3·0	—	—	· ·	1·00	
	17	76·5	—	20	—	3·0	—	—	· ·	—	0·5
III. Aluminium	19	100·0									
	20*	100·0									
	21†	100·0									
	22	—	100								
IV. Al-Zn (15 per cent Al)	11	15·0	—	85							
V. Al-Cu (4 per cent Cu)	18	96·0	· ·		—	4·0					

* Overheated to 950° C.   † Addition of Carbon.

one of the castings had, before casting, been over-heated to a temperature of 950° C, and in another case had been intentionally contaminated by the addition of carbon. For the fourth, commercial aluminum, as available at that time, containing a little over 99 per cent of aluminium, was employed. The exact compositions of the various castings used are given in Table 51.

The measurements were made with very great care by means of a large screw micrometer. This micrometer had an adjustable gap, and was set in each case by means of accurate end-gauges supplied by the Metrology Department of the Laboratory. The micrometer and these gauges were of steel, and since the coefficient of thermal expansion of steel differs markedly from that of aluminium, the temperature at which measurements were made was in every case carefully noted, and the results of measurements have been corrected for the errors arising from difference in thermal expansion between the micrometer and the castings under measurements. In every case the castings were kept in the same room as the micrometer for some hours before measurements were made in order to secure that they should attain the same temperature. Further, the micrometer itself was suspended by means of an overhead pulley and counterpoise, so that only the actual measuring ends had to be handled by the operator.

The dimensions in each case were read to $\frac{1}{10000}$ inch, and many of the determinations were checked by a second observer. Even with the greatest care, however, it was not found possible to secure consistent readings to this degree of accuracy, and the individual measurements cannot, therefore, be relied upon to an accuracy much greater than $\frac{5}{10000}$ inch. Since the actual dimensions measured were approximately as follows :—(A) about 13.2 inches, (B) about 11 inches, and (C) slightly under 10 inches, this degree of accuracy of measurement appeared to be adequate for the purpose, since a change of even $\frac{1}{10}$ of 1 per cent in the dimensions of the casting would have made itself felt very markedly in the measurements.

The results of the actual measurements are shown in the graphs of Figs. 89, 90 and 91. These graphs relate to measurements on the series of castings which had been exposed to the weather.

Fig. 80.

MONTHS AFTER CASTING

8   12   20   30   40

ries I.

0   10   20   30   90

MONTHS AFTER FIRST MEASUREMENT

1

—0

MEASUREMENT
A ————•
B - - - -o
C ·····×

Al-Zn
80/20

2

—0

80/20+Pb

3

—0

80/20+Fe

4

—0

$\frac{1''}{1000}$

80/20+Si

5

—0

80/20+Mn

6

—0

80/20+Mg

7

—0

80/20
OVERHEATED

8

—0

80/20+C

Variation in $\frac{1}{1000}$ ths of an inch (corrected to 15° C.).
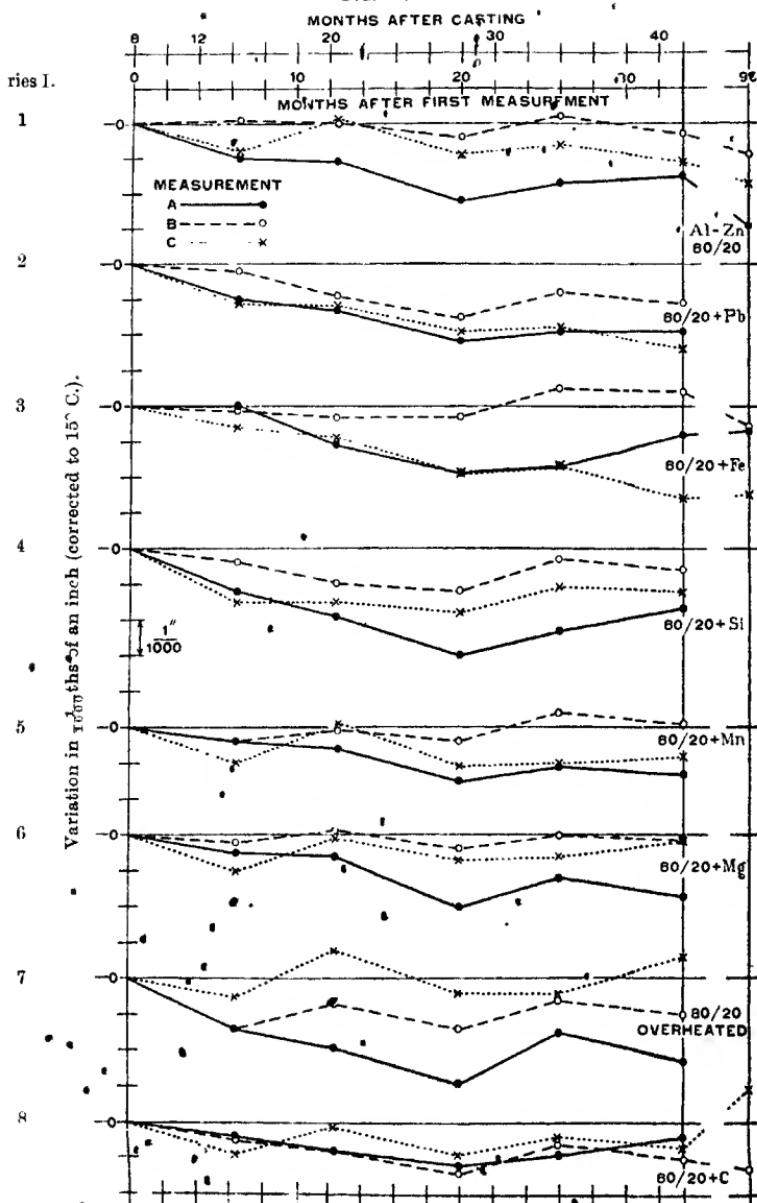
Corresponding measurements made on the castings kept in store gave results so close, similar t those represented in the Figures, that it is not thought worth whet to reproduce the second series. As will be seen from the Figures, measurements were made at approximately six months' interval over a period of about three and a half years. At the end of that period, owing to the increasing pressure of war conditions, these measurements were suspended. Unfortunately, owing to the long time required for the careful machining of the flats on the castings, the first measurements were not made until the castings were approximately eight months old. The work on the ageing of aluminium castings, however, which is described in the preceding Section of the present Report, indicates that the ageing process is by no means completed in the first eight months after the castings have been made. If, therefore, any dimensional changes are associated with the ageing phenomena, it would be confidently anticipated that such dimensional changes would continue for a period of several years after the castings were prepared, and would therefore be amply shown in the present series of measurements.

Examination of the graphs in Figs 89, 90 and 91 shows at once that actual dimensional changes of sufficient size to be of importance for practical purposes have not occurred in any of these castings. This result is interesting and important as definitely establishing the dimensional stability of castings of these kinds, even in the presence of intentionally added impurities and after comparatively serious maltreatment of the metal in the case of pure aluminium. The largest dimensional change shown on any of the graphs of the present series occurs in the case of a pure aluminium casting, and even in that case only amounts to an increase of $\frac{3}{1000}$ inch on a dimension of 11 inches. This corresponds to a change of less than 0.03 per cent in the dimension as measured. Even this minute change appears to be exceptional, and is probably due to the fact that the material is extremely soft and that the casting may have been slightly distorted during manipulation. Examination of the details of the various graphs, however, shows that certain very minute changes in dimension appear to occur in the ageing of these

FIG. 90.
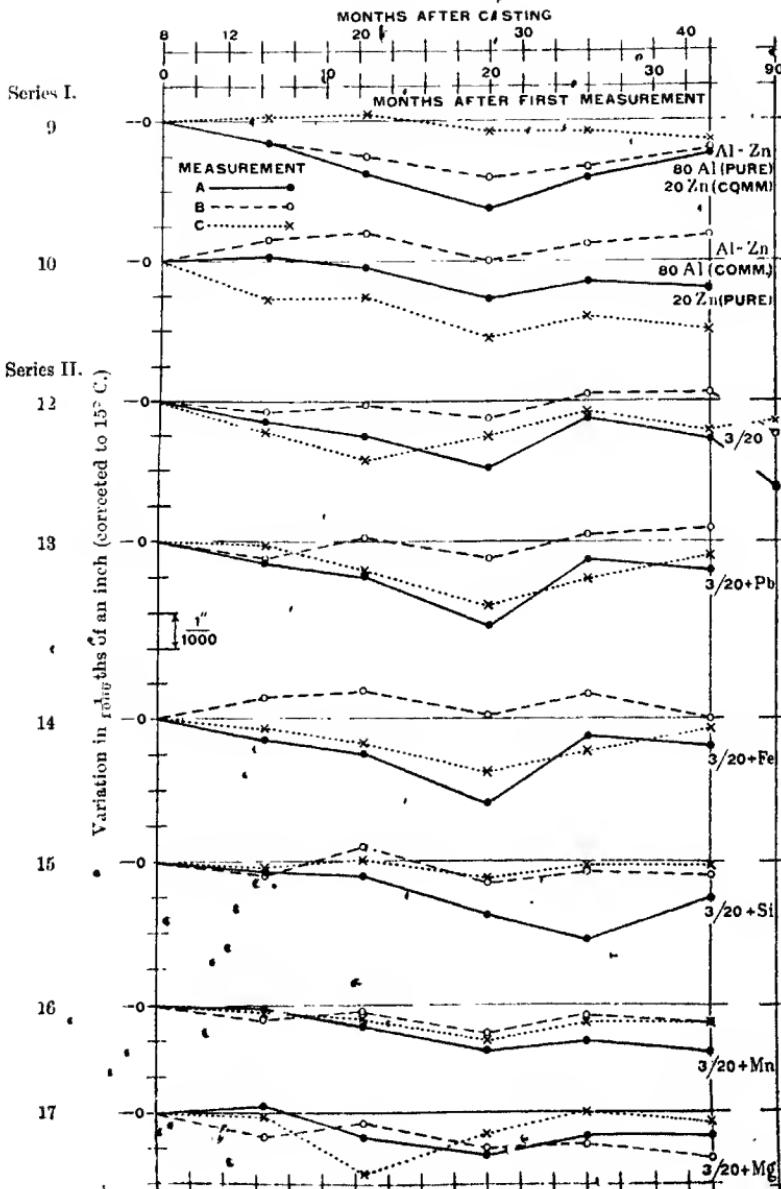
castings. Owing to their very minute nature, one is tempted to
regard them as possibly of accidental origin, arising from temperature
inequalities in the material or in the micrometer or other causes of
variation, not readily traced, or as the result of minute deformations
of the castings during handling, etc. With the exception of the
relatively large changes found in the pure aluminium casting just
referred to, and also in another casting of the same (pure aluminium)
series. however, these small dimensional changes appear to follow a
somewhat regular order, and this regularity disposes of the idea
that they can be due to accidental causes of the kind indicated
above. Taking the graph for the " A " measurement in Figs. 89
and 90 particularly, we find that this dimension in most of the
graphs, *but not in all*, reaches a minimum value in the observations
taken when the castings were twenty months old, and that after
this period this dimension appears to return to a value which in
most cases is very near that from which it started when the
measurements were first made. It is obviously very difficult to
account for any such reversal of dimensional change, but a
suggestion may, perhaps, be hazarded. It is known that the
ageing processes, described in the preceding Section of this Report,
occur at different rates and possibly to different extents, in portions
of material of the same composition which have been cooled at
different rates, and therefore in portions of the same casting of
different thickness. The minute dimensional changes which, it
might be anticipated, would be associated with these ageing
phenomena would therefore occur at different times and to different
extents in various portions of these crank-case castings. If this
were the case, it might happen that the changes first occurring
would lead to a slight diminution in such a dimension as A, Fig. 88,
and that the subsequent ageing of the thicker parts of the casting
should restore that dimension by reversing any very minute amount
of warping which had been produced by the first changes.

Although, as has been mentioned above, the regular
remeasurements of these castings at intervals of a few months
had been suspended when the castings were forty months old, the
castings themselves were preserved and the machined flats protected

Fig. 91.
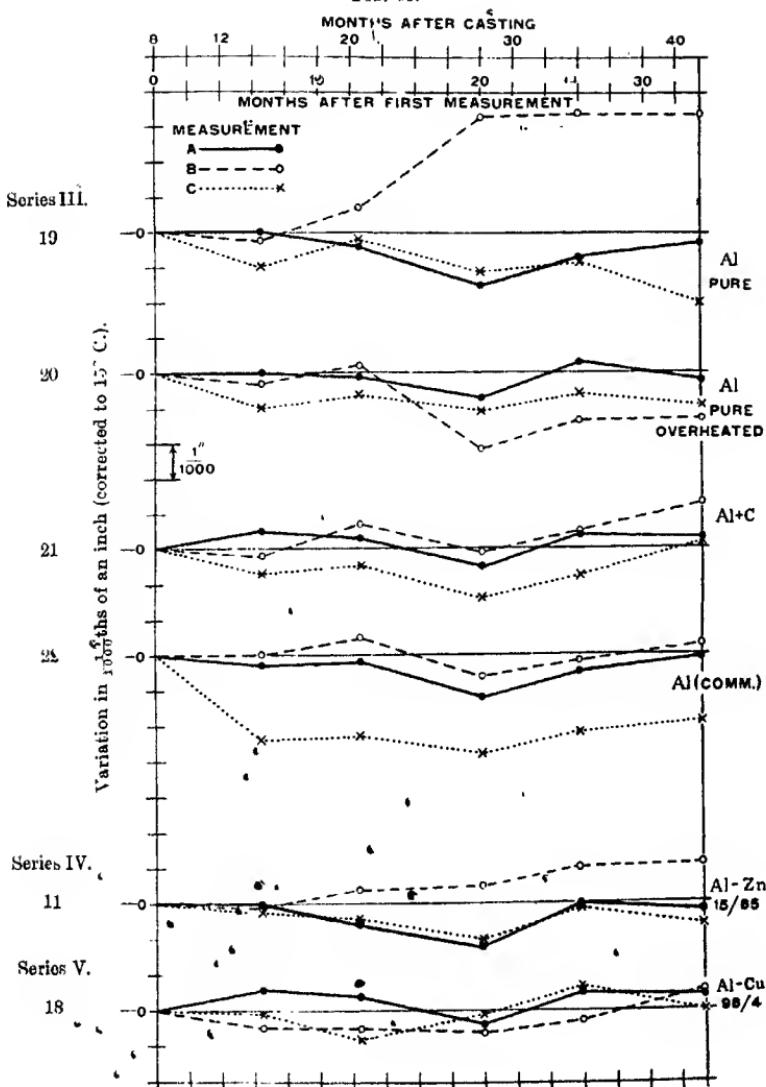
as well as possible by a thick coating of vaseline. In January, 1921, some seven and a half years (nine: months) after the production of the castings, some of them were again examined. It was found that even in the case of those which had remained exposed on the roof of the building the machined faces were apparently perfectly intact when the coating of vaseline and dirt had been carefully removed. In order to ascertain whether the castings had preserved their dimensional stability after this long exposure, four of them, namely, Nos. 1, 3, 8 and 12 were carefully remeasured in the Metrology Department of the Laboratory. Of the twelve remeasurements thus made, one only, representing dimension A of casting No. 8, gave an abnormal result, showing an apparent increase of 0·0071 inch. The two other dimensions, measured on the same casting, showed nothing unusual, and it therefore seems probable that this particular measurement has been accidentally affected by some damage to the casting, although there is no visible evidence to that effect. With the exception of this abnormal measurement, the results of these remeasurements are shown—on a broken line— at the edge of the graphs in Figs. 89 and 90. It will be seen that these show no greater departure from the horizontal straight line than is to be found in earlier parts of the same graphs, while in some cases the final departure from the original dimension is remarkably small and well within the errors of measurement. These results, therefore, as a whole, strongly confirm the conclusion that castings of these alloys, whether pure or contaminated with the elements named, possess very great dimensional stability, and that nothing in the nature of distortion—and much less disintegration—need be feared.

IV(c). *Fracture under Prolonged Loading ("Season Cracking").*

Some of the alloys described in the present Report, notably alloy " A " (3/20), and to a lesser extent, alloy " E " and its modifications " F " and " G," have been found to be liable, in certain circumstances, to a mode of fracture very similar to that which is well known in the case of brass under the somewhat misleading name of " season

cracking." Failures of this type were first noticed in the case of alloy
"A," where sheets of this material had been kept for a considerable
time closely coiled, the coils being restrained by wire. Similar
forms of fracture were also observed in certain spinnings to which
reference has already been made (see Section III (a), page 80). These
occurrences led the Authors to undertake a somewhat general
investigation into the nature of such failures, which occur in metals
when apparently exposed to stresses considerably lower than those
at which they would break if tested in the ordinary way, and
only after the lapse of a more or less considerable period of time.
The first results of this investigation were communicated by
two of the Authors in a Paper to the Royal Society.[*] This
Paper shows that such widely different metals as brass, lead, and
some aluminium alloys are capable of undergoing the same type of
fracture in analogous circumstances. This fracture is characterized
by the fact that it occurs by the pulling away of the crystals of the
metal from one another, with as nearly as possible a total avoidance
of any fracturing of the crystals themselves. Somewhat similar
types of fracture have also been observed in mild steel, as indicated
in the Paper just cited and in another Paper, also by two of the present
Authors, on " Inter-crystalline Fracture in Mild Steel," communicated
to the Iron and Steel Institute.[†]

In the first of the Papers just mentioned, a theoretical explanation
of the phenomena of failure under prolonged stress by inter-crystalline
fracture has been attempted. Although much discussion has been
devoted to an effort to disprove this theory, the Authors feel that
it still offers the only explanation yet put forward to account for
these remarkable phenomena.[‡] It does not, however, appear
desirable to enter into the theoretical aspects of the whole subject
here, since the present Report is mainly concerned with the conditions

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* " The Inter-Crystalline Fracture of Metals under Prolonged Application
of Stress." Rosenhain and Archbutt. Proc. Roy. Soc. A., Vol. 96, 1919.

† Rosenhain and Hanson. " Inter Crystalline Fracture of Mild Steel."
Journ. Iron and Steel Institute, 1920, No. II, Vol. cii.

‡ " General Discussion on the Failure of Metals under Prolonged Stress."
Faraday Society and this Institute, 1921.

in which various alloys may be liable to undergo failure of this kind, and the metho', by which the occurrence of such failures can be more or less completely prevented.

Alloy " A " (3/20).

In one of the Papers referred to above,* two of the present Authors have pointed out that the liability to inter-crystalline fracture under the prolonged application of stress is associated in the case of alloy " A " with a particular type of microstructure. This conclusion has since been very fully confirmed. A very large number of samples of the alloy have been subjected to various kinds of heat and mechanical treatment, and have then been tested for liability to fracture of this type by means of a comparatively rough but simple and effective test. The test in question, which is applicable only to metals in the form of comparatively thin sheet, consists in bending a strip in the form of the letter U, the radius of the bend being about 1½ inch. The test-piece is gently bent into this form, and the two ends are then held in position by means of a clip. In this way the metal in the bent portion of the strip is kept under a bending stress, the amount of which depends, among other things, upon the radius of curvature, and the actual pull exerted by the wire clip upon the ends. Care must, of course, be taken not to overbend the test-piece when preparing it, the necessary precaution being that the two ends of the test-piece shall never be brought appreciably nearer together than the position in which they are finally clamped. The actual stress developed in the test-pieces in the outer layers of the bent portion, is not very readily calculable. The reason is that, in these bent test-pieces, plastic deformation of the metal has taken place, and consequently the elastic limit (of restitution) has been raised considerably above that existing in the unstrained metal. It is obvious, however, that the stress existing in the bent test piece can nowhere exceed the elastic limit

* " The Inter-Crystalline Fracture of Metals under Prolonged Application of Stress." Rosenhain and Archbutt. Proc. Roy. Soc. A., Vol. 96, 1919.

of the material in the particular condition in which it exists at that time, but it is important to note that this elastic limit can never very closely approach the ultimate strength of the metal, unless the bend has been made so sharp that any even slight additional bending would lead to the immediate fracture of the piece. It is only when the ductility of a metal has been completely exhausted by overstrain, that the elastic limit is raised to the vicinity of the breaking stress. This consideration is important in view of the suggestion which has been put forward elsewhere * that the failure of brass, particularly under internal stresses existing in spun or drawn material, may really be due to the fact that the internal stresses are equal, or very nearly equal, to the normal ultimate strength.

In view of the fact just pointed out, the bend test described above must be regarded as being of a very severe nature. Careful experiment has shown that fracture of the kind now under discussion does not occur in a bent strip if the amount of bending has been sufficiently small to allow the strip to resume its original straightness after the removal of stress. This shows that fracture under prolonged loading will not occur in this alloy unless the applied stress considerably exceeds the primitive elastic limit. When material is used in practical construction, care is, or should be, taken to exclude the occurrence of working stresses too closely approaching the elastic limit, while stresses exceeding the elastic limit are not ordinarily regarded as admissible. It would seem, therefore, that under all normal service conditions, the material could not be exposed to prolonged loading with intensity sufficient to produce fracture of this kind. Stresses which are at once very intense and are continuous in their application, only arise where these are produced in the material by severe cold working. Where material is severely locally strained by such operations as cold stamping, sharp bending, punching or shearing, or similar operations, conducted in such a manner as to leave it severely internally stressed, failure in the course of time may result if the material is used in the special

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* Hatfield and Thirkell. Journ. Inst. of Metals, 1919, No. II.

condition which, as will be seen below, is necessary to render it sensitive in this direction.

The indications of the bend test which has now been studied for several years, are very fully borne out by the behaviour of the alloys in practical use. The behaviour of spinnings, stampings and of actual riveted girders constructed out of channels and stampings of alloy "A" (3/20) entirely bears out the indications of the bend test. It is found that material which would stand this test for a period of several days, never undergoes failure under ordinary practical conditions, at all events within a period of some two or three years

In the case of alloy "A" (3/20), indications of the bend test show in the first place that the alloy as produced in the ordinary way by hot rolling, is entirely free from risk of inter-crystalline fracture under prolonged loading. This is true, at all events, of material which has been produced in the manner adopted in more recent practice, where temperatures for rolling and intermediate heating have been intentionally kept somewhat lower than was originally done. Even if a rolling temperature as high as $450°$ C. has been used, however, the bend test shows definitely that, provided a subsequent reduction by rolling of not less than 50 per cent of the cross-sectional area has been applied, the material is still entirely safe. With so high a rolling temperature as $450°$ C., and the use of smaller amounts of reduction by hot working, the material is found to be liable to fail under the bend test if exposed to it for a considerable time. More interesting indications are obtained when the bend test is applied to material which, after final rolling, has been annealed at various temperatures. With an annealing period of one hour, it is found that material annealed at temperatures from $400°$ C. to $450°$ C. undergoes failure by the bend test in about one hour. If annealed at somewhat lower temperatures, ($350°$ to $400°$ C.) it is found that failure under the bend test will occur in a somewhat longer time varying from one to four or five hours. With still lower temperatures ($300°$ to $350°$ C.), considerably longer exposure to the bend test is required to bring about failure ; in some cases periods as long as twelve hours are required. With material which has been annealed between $250°$ and $300°$ C., bend tests have in some cases shown

completely satisfactory resistance over a period of several years;
in other cases, however, fracture has occurred after a few weeks'
or several months' exposure. On the other hand, material which
has been annealed at a temperature not exceeding 250° C. is found
to withstand the bend test indefinitely.*

It has already been indicated that the type of fracture occurring
in these circumstances is peculiar, in that it takes place by the
pulling apart of the crystals and not by the fracture of the crystals
themselves, thus producing what is known as a typical inter-crystalline
fracture.   It is not surprising, therefore, to find that the liability
to undergo failure of this kind varies with the nature of the crystal
boundaries in the particular specimen of metal under test.
Microscopic observation at once shows that rapid failure occurring
in an hour or so under the bend test is associated with a completely
equi-axed microstructure accompanied by very smooth regular
crystal boundaries.   Such a structure, together with a typical crack,
and clearly showing the inter-crystalline nature of the fracture,
is shown in Fig. 92, Plate 15, under a magnification of 150 diameters.
The structure there illustrated is typical of alloy " A " (3/20), which
has been annealed at 450° C.   It is interesting to remember in this
connexion that annealing at 450° C. does not produce a satisfactory
softening of this alloy, which can be obtained in a much softer
condition by annealing at low temperatures such as 200° C. or
250° C.   Indeed, material which has been rendered very soft and
ductile by the latter low temperature annealing is again rendered
harder and less ductile by heating to 450° C., whether followed by
rapid or slow cooling.

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* Since the above was written, a Paper by Moore, Beckinsale and Mallinson
—Journ. Inst. of Metals, 1921, No. I—has drawn the authors attention to the
part which chemical action on the inter-crystalline material may play in
connexion with " Season Cracking."   They accordingly tested the behaviour of
Alloy " A " (3/20) under the bend-test when entirely protected from chemical
action by being placed either in a good vacuum or in an atmosphere of pure
hydrogen.   In both cases, fracture was considerably retarded but never
completely prevented.   An account of these experiments has been given by
W. Rosenhain in his introductory address to the General Discussion on
" Season Cracking " already mentioned.

The use of lower annealing temperatures than 450° C. does not result in so rapid or so complete a production of a completely equi-axed structure. The safe annealing temperatures below 250° C., in fact, do not produce any serious amount of equi-axing even in several hours. The material annealed at these temperatures, although rendered very soft and ductile, retains an elongated microstructure. This is a remarkable instance of a phenomenon observed in other materials, where it is frequently found that mechanical softening precedes the formation of a new equi-axed microstructure. In most materials, however, this stage in which the metal is soft without being equi-axed is transitory. In the case of the alloy "A" (3/20), on the other hand, it can be retained permanently, and in fact leaves the material in the best and soundest condition. The typical microstructure of the alloy "A" after annealing at 250° C., and therefore in a condition in which it will not undergo inter-crystalline fracture under prolonged loading, is illustrated in Fig. 33, Plate 15, under a magnification of 150 diameters.

In view of the results obtained from the systematic employment of the bend test which has been described above, the Authors regard material which has been heated or annealed at any temperature above 250° C. as being in an " over-annealed " condition. It is only in this over-annealed condition, and if left without further treatment, that the alloy is liable to undergo intercrystalline fracture as the result of prolonged and severe loading. The next step in the Authors' investigation of this subject consisted in an endeavour to ascertain by what means, if any, over-annealed material could be restored to a safe condition from the present point of view. The treatments first tried were purely thermal, consisting in re-heating the over-annealed alloy to various lower temperatures, followed either by slow cooling or by quenching. The results obtained were entirely negative. Material which has once been over-annealed cannot be restored to a safe condition by any subsequent heat treatment at lower or at higher temperatures. The application of mechanical work on the other hand proved to have a very powerful influence on the phenomena.

The application of cold work to material over-annealed by

heating to 450°C. was found considerably to modify the behaviour under the bend test. Various amounts of reduction by cold-rolling were tried, and it was found that increasing distortion of equi-axed crystals in the over-annealed alloy tended materially to retard the rate of failure under the bend test. The results indicate that material which has not been excessively over-annealed could probably be rendered completely safe by a relatively small amount of cold working. On the other hand, seriously over-annealed material cannot be prevented from ultimately failing under the bend test by cold working alone. It must be remembered, however, that from the point of view of other physical properties of the material, the application of cold work to the alloy "A" is in itself undesirable. The effect of hot-rolling upon previously over-annealed material was next investigated. The results of a long series of experiments can be summarized in the statement that material which has been over-annealed at 450° C. can be rendered completely immune from failure under the bend test, by hot rolling it to a sufficient extent. A total reduction of not less than 50 per cent of the cross-sectional area of the material if not too rapidly applied, is found to be sufficient for this purpose. Microscopic examination shews that this mechanical treatment completely changes the original equi-axed crystals having smooth boundaries, and replaces them by elongated grains whose boundaries are much more irregular in character.

In general terms, the outcome of the investigation briefly outlined above, on the behaviour of alloy "A" (3/20) under the bend test, after various forms of thermal and mechanical treatment, may be summarized as indicating that, provided the temperatures employed and the amount of mechanical reduction applied are satisfactorily and carefully regulated, the alloy can be produced in a condition in which it is free from all risk of undergoing inter-crystalline fracture as the result of prolonged exposure to severe loading.

As throwing a considerable amount of further light upon the nature of the process by which inter-crystalline fracture occurs in this alloy, a series of tests may be quoted in which fracture of this type has been produced by direct tensile loading in an ordinary

testing-machine. The results of two series of tests are tabulated in Table 52 relating to material of similar composition but in different states as regards annealing and mechanical treatment.

It will be seen that in both series of tests there is a very considerable time effect. Material which requires in one instance 25·4 tons per square inch to produce fracture in 6 minutes, undergoes fracture in 50 minutes under a load of only 21 tons per square inch.

TABLE 52.

Tensile stress per square inch.	Time taken to produce rupture.
Tons.	Mins.
Series I.	
16	9
14	31
13	56
Series II.	
21	50·0
22	65·0
23	27·0
24	28·5
25	13·0
25·4	6·0

The existence of the time effect shown in the above Table, and the manner in which the time increases rapidly with decrease of the applied load, suggests that the phenomenon is one of viscous flow, differing entirely in character from the so-called "flow" which occurs during the ordinary plastic deformation of a piece of metal under tensile overstrain. The occurrence of such a time effect, and its strong suggestion of viscous flow, is in close agreement with the theoretical explanation of the phenomena of inter-crystalline fracture

under prolonged stress which has been put forward by two of the
Authors in the Paper already referred to.

If the view is correct that the occurrence of inter-crystalline
fracture of the kind here under discussion depends largely upon the
physical properties of an inter-crystalline cement (mainly its viscosity)
and upon the shape and arrangement of inter-crystalline boundaries,
it would naturally be anticipated that variations in composition
would materially affect the behaviour of a given alloy. It was
thought that certain additions to the alloy "A," for instance, might
materially alter the hardness and viscosity of the inter-crystalline
cement, or affect the manner in which rearrangement of crystal
boundaries would occur. The influence of various additions to
the alloy was, therefore, investigated, the additions employed being
(1) silicon, (2) nickel and (3) manganese.

Additions of silicon up to 1 per cent were tried, but were found
to be substantially without effect upon the behaviour of the alloy
under the bend test. Examination of the microstructure also showed
that the additional silicon did not materially affect the behaviour
of the alloy during annealing. The addition of nickel, on the other
hand, was found to produce a marked but not a very large effect.
Material containing $\frac{1}{2}$ per cent of nickel in addition to the 3 per
cent of copper and 20 per cent of zinc normally present in the alloy,
was found still to undergo cracking under the bend test if annealing
had been carried out at 450° C. It was found, however, that instead
of producing fracture in about one hour, the bend test required a
period of several hours before fracture occurred. This marked
retardation suggests that in the presence of $\frac{1}{2}$ per cent of nickel, a
somewhat higher annealing temperature than 250° C. could be
used with safety, or, what is really more important, the limit of
250° C. for the safety of the material need not be so strictly observed.

In order further to test the influence of nickel on alloys of this type,
a material was prepared in which the whole of the copper ordinarily
contained in alloy "A" was replaced by nickel, thus giving an alloy
containing 20 per cent of zinc, 3 per cent of nickel, remainder
aluminium.. Ordinary tensile tests on this material give results
distinctly lower than those obtained from the alloy "A," and its

behaviour under the bend test was found to be no better than that
of the alloy of normal composition. It is possible that the combination
of higher nickel and copper contents together might produce
improved results, but at the time when these experiments were
carried out the Authors had not yet discovered the fact that
the addition of nickel to an alloy already somewhat hard, could
be made without further increasing its hardness. These higher
compositions, therefore, have not been tested.

A much more marked effect was produced by the addition of
$\frac{1}{2}$ per cent of manganese to the alloy "A." Unfortunately, the
addition of $\frac{1}{2}$ per cent of manganese makes the production of the
alloy by forging and rolling distinctly more difficult than that of
the normal material. These difficulties can, however, be overcome
by suitable careful treatment. The effect of manganese upon the
behaviour of the alloy during annealing, and under the bend tests
is very marked indeed. It has been found that in the presence of
$\frac{1}{2}$ per cent of manganese, several hours' exposure to a temperature
of 450° C. does not result in complete equi-axing. The microstructure
of rolled sheet of this alloy annealed for 6 hours at 450° C. is shown
in Fig. 94, Plate 15, under a magnification of 150 diameters. It was
anticipated from this observation, that alloy "A" containing $\frac{1}{2}$ per cent
of manganese would withstand the bend test indefinitely, even after
annealing for one hour at 450° C. Actual experiments verified this
anticipation to a very considerable extent. A number of bend test-
pieces treated in the ordinary manner as described above, were
found intact after two years' exposure. In order to carry the matter
further, and to render the test still more severe, these bend test-
pieces, which had successfully withstood the ordinary tests for so
long a time, were further stressed by pressing the ends more tightly
together. In one instance only such a bend test-piece, after
restressing in this manner, broke in the course of a few hours. This
intensification of the test after such very prolonged exposure, makes
it unduly severe in comparison with the test ordinarily applied to
alloy "A" and to other materials, but it serves to show that the
possibility of undergoing fracture by inter-crystalline cracking
resulting from prolonged exposure to severe stresses, has not been

entirely eliminated in material annealed at 450° C. by the presence
of a small amount of manganese, although obviously the power
of the material to resist failure of this type has been enormously
increased.   There can be no doubt, however, that material which
has been less severely over-annealed than that referred to in the
experiments just described, will be entirely immune from failure
of this type if it contains ½ per cent of manganese.

The influence of manganese in retarding the annealing, or rather
the recrystallization, of alloy "A" is of somewhat general metallurgical
interest.   It has been shown that the addition of a small amount of
manganese to certain copper alloys produces a similar effect in
retarding the rate of crystal growth (*).   The theoretical explanation
of this influence of manganese, however, need not be considered
here.

In connexion with the very marked influence of manganese
on the phenomena of inter-crystalline fracture, it occurred to the
Authors that there was a possibility that the real effect of over-
annealing might reside in some chemical action of the surrounding
atmosphere upon the inter-crystalline cement in the alloy during
annealing, and that this action might be eliminated or modified by the
presence of manganese.   In order to test this point, a series of samples
of alloy "A" (not containing manganese) were annealed in special
atmospheres, consisting of hydrogen and nitrogen respectively, thus
avoiding all possibility of atmospheric oxidation.   The material
used for these experiments was rolled sheet giving in its original
finished condition satisfactory resistance under the bend test.
Samples of such sheet were then annealed for one hour at 450° C.
in precisely the same manner as those which had previously been
annealed in air, as described above.   It was found that samples
annealed in these special atmospheres failed in precisely the same
manner as if they had been annealed in air, the annealing in both
cases having produced complete equi-axing of the microstructure.
This experiment conclusively disposes of the possibility that inter-

* Rosenhain and Hanson.   "The Properties of Some Copper Alloys."
Journ. Inst. of Metals.   1919, No. 1, Vol. xxi.

crystalline oxidation, during annealing plays any part in the phenomena discussed in the present section.

Alloy " E " and its Modifications.

These alloys, being very similar in constitution and composition to alloy " A," containing very similar amounts of zinc and copper, but with additions of manganese and magnesium, were naturally suspected of being capable of undergoing the same type of failure under prolonged stress as the alloy " A." It was hoped, however, that the presence of manganese would, as in the case of alloy "A" itself, substantially retard or even entirely prevent the occurrence of failure of this type. In the case of these alloys, which for purposes of hardening have to be heated and quenched after their final mechanical treatment, the question of the safe maximum temperature which can be employed is one of particular importance.

In applying the bend test in the same form as that which had previously been adopted for alloy " A," to alloy " G," and its modifications " E " and " F," it must be borne in mind that the very much higher primitive elastic limit of these alloys renders the bend test correspondingly more severe, because it allows of the retention in the material of much higher unit stresses than those which can be produced in alloy " A." None the less, it has been found that failure under the bend test in the case of alloys " E," " F " and " G," occurs very much less readily than in the case of alloy " A "

Since it has been found that a temperature of 400° C. is sufficient, by means of quenching and ageing, to develop the best physical properties in the material so far as ordinary tensile tests are concerned, the effect of higher temperatures for heat treatment has not been further investigated When quenched from 400° C., the behaviour of these alloys varies according to the intensity of stress to which they are exposed. Under the bend test as applied at the National Physical Laboratory, a number of samples of the alloys thus treated have failed after more or less extended exposure, giving in some cases a typical inter-crystalline fracture such as that found in alloy " A," and in other cases failing by a process of splitting

which sugg....... .. .. what different mode of rupture, depending upon a certain degree of lamination in the material. Similarly treated material which has been tested by the bend test at the Royal Airship Works, Cardington, has in no instance failed. Apart from possible small differences in the severity of the bend test employed, it appears probable that the exact method of manufacture, and particularly the amount of reduction applied during the last stages of hot-rolling, affect these results. If, however, the temperature for quenching is reduced to 375° C. or still more certainly if it is reduced to 350° C., the failure of these alloys under the bend test, even in its most severe form, is eliminated and the material rendered completely immune from inter-crystalline fracture of the type under consideration.

The indication of the bend test in these alloys has been very fully borne out by the behaviour of the same materials when employed for construction. At the Royal Airship Works at Cardington, complete airship girders have been constructed of alloy " E," and neither in the stampings nor around the rivets of these girders have there been any signs of cracking, even in samples more than a year old. This statement applies even to material which has been quenched from 400° C., thus indicating that the bend test, as employed at the National Physical Laboratory, imposes conditions of much greater severity than those likely to be met with in practical construction.

The satisfactory behaviour of alloys " G," " E " and " F," when treated at or below 350° C. under the bend test and in practice, serves to confirm the view expressed above as to the value of an addition of manganese in preventing failure of the type under discussion. An examination of the microstructure of these alloys in the condition which they acquire after heat treatment, further confirms this view. This microstructure is illustrated in Fig. 95, Plate 15, under a magnification of 150 diameters. It will be seen that in spite of having been heated for 30 minutes at 400° C, the material is by no means equi-axed, a result which is undoubtedly due to the retarding influence of manganese on recrystallization.

Alloy " Y."

A very large number of test-pieces of this alloy have been subjected to the bend test, but in no instance has cracking occurred. So far as the investigations of the Authors have yet gone, this alloy appears to be entirely immune from cracking under the effect of prolonged loading. The behaviour of the material in actual construction fully bears out this result. A similar series of tests made with alloys of the " D " type (resembling Duralumin) have also given similar negative results under the bend test. In the case of these latter alloys, however, certain observations made on actual airship structures suggest that occasional instances of cracking after considerable time have occurred. The exact causes and nature of these cases have, however, not yet been thoroughly investigated by the Authors.

IV(d). Corrosion.

A knowledge of the corrodibility of aluminium alloys under various conditions is undoubtedly of the greatest possible importance. The Authors, however, wish to emphasize the view that the corrodibility of material should be carefully studied in regard to the particular condition in which it is to be used, or, conversely, that the use of any particular material should be regulated in such a manner as to avoid the conditions under which it may be particularly liable to corrosion. The extreme view sometimes adopted, that an alloy liable to severe corrosion under a particular condition should not be used at all, is, in the Authors' opinion, unjustifiable, since it would lead to the general abandonment of alloys which in certain conditions are capable of rendering most valuable service.

In the earlier stages of the research described in the present Report, the systematic study of the corrodibility of alloys was regarded as an important part of the work. In the later stages under war conditions, when it had been recognized that there could be little or no doubt as to the relative corrodibility of the various types of alloys, the further detailed study of corrosion was abandoned temporarily in favour of the search for information

more immediately required. In the final stages of the work, the study of corrosion has not been resumed in detail in view of the fact that a special Committee for the investigation of the corrosion of aluminium alloys had been set up by the Department of Scientific and Industrial Research, and that later its functions had been transferred to a sub-committee of the Corrosion Research Committee of the Institute of Metals. The fact that by this organization the research on the corrosion of aluminium alloys has been placed in particularly competent and experienced hands, and that, moreover, one of the present Authors is a member of the sub-Committee in question, has led them to discontinue further detailed work on corrosion. In the present Report, therefore, some of the data obtained in the earlier stages of the research are reproduced, together with general conclusions as to the relative corrodibility of the various alloys described, which have been derived from the Authors' observation and experience both in their own handling of the alloys and in various experiments, and from the observations and experience obtained with them in actual use, particularly at the Royal Airship Works, Cardington.

Even before the discussion on the Tenth Report to the Alloys Research Committee, it had been recognized that the methods formerly adopted, of estimating the corrodibility of any metal by measurement of the loss of weight after prolonged exposure to corrosive conditions, such as immersion in sea-water or in tap-water, led to unsatisfactory results. It was realized that local attack, such as pitting, might practically destroy a piece of metal while the total loss of weight remained very small. In view of this consideration, it was at that time decided to adopt an entirely different method of studying corrosion. This consisted in the preparation of a series of tensile test-pieces, machined and finished ready for testing, and their exposure to the corrosive conditions to be studied. From time to time, at successive stages of exposure, individual members of these series of test-pieces are removed and tested in the tensile testing-machine. Provided that a reasonably uniform material is available, such tensile tests should show, in the absence of corrosion, very little or no change in the results. The

occurrence of even small amounts of pitting, however, would immediately reduce both the ultimate strength and the elongation of the material very appreciably. In this way it was hoped to show the relative resistance to pitting or local action in various alloys.

In accordance with this scheme, a series of test-pieces representing five alloys of the aluminium-zinc system, and one ternary alloy of aluminium, copper and zinc, in the form of hot-rolled rods, were prepared. The material used was annealed rolled rod, ½ inch diameter, which was cut into 6 inch lengths. The middle

TABLE 53.

| Alloy No. | Composition by analysis. | |
	Zn.	Cu.
	Per cent.	Per cent.
D 09 B . . .	9·25	—
D 12 . . .	13·24	—
D 15 B . . .	15·05	—
D 20 B . . .	20·15	—
D 25 BB . .	24·50	—
H 4 A . . .	20·06	2·95

portion of each piece was then turned down over a length of 3 inches to a diameter of 0·356 inch, the turned portion being kept concentric with the unmachined ends. The preliminary tests having shown that the material in question was sufficiently uniform, the test-pieces were loosely mounted in a teak frame and totally immersed in the sea at Portsmouth Dockyard at the end of May 1912. This frame, ready for immersion in the sea, is shown in the photograph, Fig. 96, Plate 15. The frame contains six different alloys, eight test-pieces of each being provided. The compositions, by analysis, of the six sets of bars contained in this frame are shown in Table 53.

Specimens for re-testing were removed at the end of 94, 206, 314, and 448 days. The results of these tests are shown in Table 51, which refers to the binary alloys of aluminium and zinc only.

Inspection of this Table shows the remarkable result that there has, on the whole, been no appreciable deterioration in tensile strength of these test-pieces by corrosion, even after 448 days in the sea. The very slight diminution in the elongation figures is to be ascribed to the fact that the elongation on the unexposed test-pieces was measured on too short a length (1 inch gauge length), while those on the re-tested pieces were measured on 2 inches. It is interesting to note that in several cases there is at first a slight increase in the ultimate strength of the alloys. In the light of the subsequent study of the ageing properties of these materials, it is not surprising to find that such an improvement has occurred. In the absence of exposure to the sea it is probable that this improvement would have been possibly somewhat more marked, but the difference appears to be due rather to the relatively low temperature at which the materials were kept by immersion in the sea, than to any corrosive action. This view is fully borne out by an examination of the actual test-pieces after exposure. On the parallel machined portions of these test-pieces there is not the slightest sign of corrosion having occurred, the surfaces being as bright and smooth as when originally immersed. It is an interesting and important fact, however, that the thicker ends which had not been machined, and which were partly in contact with the teak wood of the frame, were appreciably corroded. It is this very wide difference in behaviour according to the exact character of the surface, and the manner of exposure, that probably accounts for the fact that the remarkably good behaviour of these binary aluminium-zinc alloys in the series of tests just described, has not been entirely confirmed by the work of other investigators using different conditions of exposure.

The degree of protection against corrosion when totally immersed in sea water, which seems to have been afforded to the binary alloys of aluminium and zinc by the preparation of a bright machined

TABLE 54.

Alloy No.	—	Before immersion.	Immersed in sea. (Days.)			
			94.	206.	314.	448.
D 09 B 9 per cent Zn.	Diameter	0·356	No change.			
	Yield stress (Tons per sq. in.)	4·30	4·85	4·40	5·65	3·70
	Ultimate stress. (Tons per sq. in.)	9·10	9·75	11·00	13·60	9·20
	Elongation per cent on 2 in.	43·0*	29·0*	29·0	9·0†	30·0
D 12 12 per cent Zn.	Diameter	0·356	No change.			
	Yield stress (Tons per sq. in.)	4·65	5·40	5·30	6·65	6·25
	Ultimate stress. (Tons per sq. in.)	11·10	12·40	12·90	12·90	12·80
	Elongation per cent on 2 in.	36·0*	27·5	30·0	31·0	27·5
D 15 B 15 per cent Zn.	Diameter	0·356	No change.			
	Yield stress (Tons per sq. in.)	7·70	9·00	11·25	12·10	12·00
	Ultimate stress. (Tons per sq. in.)	15·25	17·35	18·05	17·75	17·75
	Elongation per cent on 2 in.	30·0*	19·0	21·0	21·0	17·5
D 20 B 20 per cent Zn.	Diameter	0·356	No change.			
	Yield stress (Tons per sq. in.)	15·25	16·25	11·60	16·50	15·05
	Ultimate stress. (Tons per sq. in.)	20·50	21·15	20·40	20·60	20·10
	Elongation per cent on 2 in.	23·0*	14·0	17·0	15·0	15·2
D 25 BB 25 per cent Zn.	Diameter	0·356	No change.			
	Yield stress (Tons per sq. in.)	16·70	17·25	16·40	17·35	17·35
	Ultimate stress. (Tons per sq. in.)	20·40	18·90	18·85	18·80	18·75
	Elongation per cent on 2 in.	21·0*	17·5	15·0	16·0	13·0†

* Elongation measured on 1 inch, remainder on 2 inches.

† Fracture occurred at gauge mark.

surface, does not appear to be effective in the case of the ternary
alloys of aluminium with zinc and copper. A series of test-pieces
similar to those described above were prepared of a number of
these ternary alloys, and these also were immersed in the sea. At
the end of the first period of exposure, however, it was found that
these test-pieces had corroded so badly that they could be broken
with the fingers, and no tensile tests on them could be carried out.
This experiment at once demonstrated the fact, which has since
been very fully corroborated, that the ternary alloys of aluminium
with copper and zinc are very readily corroded by exposure to sea
water, and there is evidence to show that increase both of copper
and of zinc content tends to raise the degree of corrodibility. On
the other hand, it must be clearly understood that this extensive
corrodibility arises only when the alloys are *directly* exposed to sea
water, and in a lesser degree when they are immersed in tap water
or other less active solutions. When merely exposed to the air, and
particularly when not exposed directly to the weather, they preserve
a bright polished surface almost indefinitely. The Authors have
used various utensils, such as rulers, paper weights, lamp stands, etc.,
made of alloy "A" and of various other ternary copper-zinc-aluminium
alloys, for a considerable number of years in their offices. These
show not the slightest sign of deterioration in spite of the fact
that they are daily handled and are only very rarely cleaned or
polished. The relative corrodibility of these alloys when exposed
to water, therefore, must not be taken as a measure of their power
of resisting oxidation or tarnish in the air. In view of the
very great progress which has been made in the preparation of
varnishes and other protective coatings for light alloys, this degree of
permanence of these otherwise very valuable materials when
protected from the direct action of corrosive solutions, is a matter
of very considerable importance. Where these materials are not
exposed directly to contact with water, but merely to damp air
or to occasional spray, covering by a good protective coating could,
it is believed, be relied upon to preserve them indefinitely.

For the reasons indicated above, quantitative data in regard to
corrosion of other alloys referred to in the present Report, such as

alloys " E " " F," and " G," alloy " Y " and alloys of the " D " type
(resembling Duralumin, are not available. On the other hand, a large
amount of experience of the behaviour of these alloys has been
obtained, and the Authors, pending more fully detailed investigation,
would put forward the following provisional conclusions as to the
relative powers of various alloys to resist corrosion.

(1) *Alloy " Y."*—Both in the cast and in the rolled condition, this
alloy possesses considerably better powers of resisting corrosion, even
when exposed to sea water, than any of the other alloys described
in this Report. In comparison with alloys " A," "E," " F " and " G "
the difference in the powers of resisting corrosion is so great that
quantitative data are scarcely necessary ; it is sufficient to observe
the behaviour of the two materials for a short time under almost
any exposure to severely corrosive conditions. Comparison with
alloys of the " D " type (resembling Duralumin), and with Duralumin
itself, requires more careful consideration, as the difference is not
so strongly marked ; in several instances, however, which have come
under the Authors' observation there could be no doubt as to the
better behaviour of alloy " Y." In view of the fact that this alloy
contains a considerable proportion of magnesium, this result is,
perhaps, somewhat surprising. It must, however, be correlated
with what is already known as to the constitution of this alloy, to
the effect that particularly in the heat-treated condition, it does
not contain free aluminium-copper compound ($CuAl_2$).

(2) *Alloys " A," " E," " F " and " G."*—These materials cannot
be exposed to contact with sea water, or even prolonged contact with
weaker saline solutions, such as tap water, for any considerable length
of time without undergoing very serious corrosion, ultimately leading
to complete disintegration. Careful observation with the microscope
has shown that corrosion in these alloys, and particularly in alloy " A "
follows the intercrystalline boundaries, but in rolled material it
shows a preference for boundaries running parallel to the direction
of rolling. This gives rise to a condition of lamination or exfoliation
which is peculiar to these materials. When adequately protected

by a varnish or other coating, or when kept in an ordinary room where they are protected from the direct action of the weather, these materials retain a bright surface for a very prolonged period. They do not appear to be subject to atmospheric corrosion to any marked extent.

(3) *Other Casting Alloys.*—Such alloys as "L5" ($2\frac{1}{2}$ per cent copper, $12\frac{1}{2}$ per cent zinc) and alloys containing relatively high percentages of copper, such as the alloy known as "L8," containing 12 per cent of copper, are distinctly more corrodible than castings of alloy "Y." Of alloys of this type, "L5," containing relatively little copper and zinc, appears to be distinctly better than those containing higher proportions of either metal alone or of both metals together.

Section V.

Constitution of the Alloys.

Introduction.

The previous Sections of the present Report have been devoted to a description of the properties and behaviour of a large number of aluminium alloys, and in order to shorten and simplify this as much as possible, references to their structure and constitution and to the theoretical knowledge underlying and guiding the various steps in their treatment and production, have been very largely omitted. It is not too much to say, however, that such success as has been attained in the development of the various groups of alloys would not have been possible without a concurrent study of the alloys from the theoretical side, and the information and guidance which that study provided. The account of the work on the constitution and structure of the alloys has, for convenience, been concentrated in the present Section of this Report, but the amount of work to be covered is so large that only an outline can be given here. In some cases fuller publication, especially of the more purely theoretical aspects of the work, has been or will be made elsewhere, and readers particularly interested in this

aspect of the subject are referred t. these separate papers for fuller details.*

The work on the constitution and structure of aluminium alloys covered by the present Report deals in the first instance with the study of the equilibrium diagrams and models relating to the alloys of the ternary system, copper-zinc-aluminium, and an account of this work occupies the first portion of the present Section. This work has been carried out in continuance of the plan already adopted in the preparation of the Tenth Report, and its completion and continuance has been necessitated by the concurrent study of methods of producing alloys of practical importance belonging to this group.

The second portion of the present Section relates to alloys of aluminium with iron and silicon. The study of this ternary system has not been carried very far, mainly because the alloys in question are not in themselves of any considerable importance or interest. It is, in fact, only the near vicinity of the aluminium corner of the ternary system with which we are concerned. Iron and silicon occur, as is well known, as inevitable impurities in all commercial aluminium, and the various forms assumed by these components are of considerable importance in the micrographic study both of pure aluminium and of its alloys.

The third portion of the present Section relates to the alloys of aluminium with magnesium and silicon. The study of this system in very considerable detail has been undertaken because some preliminary experiments showed that the two elements, silicon and magnesium, are fundamentally concerned in the important phenomena which occur when alloys in which they are present are quenched and hardened by subsequent ageing. The phenomenon of hardening by quenching and ageing in the case of aluminium

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* Hanson and Archbutt. " Micrography of Aluminium and its Alloys." Journ. Inst. Met., 1919, vol. xxi, page 291.

Hanson and Gayler. " Aluminium-Magnesium Alloys." Journ. Inst. Met., 1920, vol. xxiv, and " Constitution and Age-Hardening of the Alloys of Aluminium with Magnesium and Silicon." *Ibid.* 1921, Vol. xxvi.

198

alloys containing magnesium was, as is well known, first discovered
by Wilm * and its practical application led to the development of
the well-known proprietary group of alloys called "Duralumin."
The discoverer of the phenomenon himself, however, expressly
disclaims being able to furnish any explanation of its real nature.
Since this phenomenon plays so fundamental a part in the
production of the light alloys having the highest strength and in
many ways the most valuable properties, it became very desirable
in the course of the present research to make an attempt to
elucidate it more fully. The results of this study are embodied
in the portions of the present Section which deal with the
constitution and equilibria of the ternary system aluminium-silicon-
magnesium, and with a detailed study of the phenomena of
hardening and ageing in aluminium alloys containing these elements,
and in some cases copper, nickel, and zinc. Although these results
are here published for the first time in detail, they were arrived
at some considerable time ago and have been utilised for
confidential purposes for several years. Full publication was
undesirable under the conditions of war existing at the time when the
work was completed. At a somewhat later point the study of the same
problem was taken up by the Bureau of Standards in America and
certain conclusions were arrived at there.† To a considerable
extent the explanation of the hardening of Duralumin and similar
alloys suggested by the American workers is in principle similar to
that arrived at by the present Authors, but the American workers
attribute the hardening effect to the compound $CuAl_2$, whereas it
is shown in the present work that this compound is not able to
produce hardening of the order of that found in the heat treatment
of these alloys, and that the part corresponding somewhat to that
played by cementite (Fe_3C) in carbon steels is played in these alloys
by the compound Mg_2Si.

* Wilm. Metallurgie, 1911, vol. viii, page 225.

† Merica, Weltenberg and Freeman. Bulletin of the Bureau of Standards,
1919, vol. xv.

Merica, Waltenberg and Scott. Bulletin of the Bureau of Standards,
1919, vol. xv.

V(*a*). THE CONSTITUTION OF THE ALLOYS OF ALUMINIUM WITH
ZINC AND COPPER.

The equilibrium diagrams of the binary systems of which this
ternary system is built up are known with some considerable degree
of accuracy in their main outlines and form the subject of previous
reports to the Alloys Research Committee. Certain features of
these diagrams, relating to the constitution of the alloys in the
solid state, have, however, not been fully determined in previous

FIG. 97.
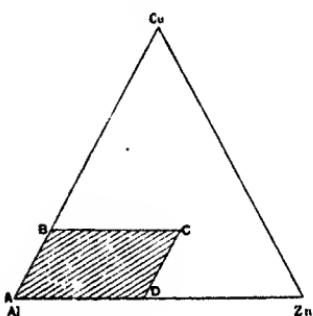
FIG. 98.
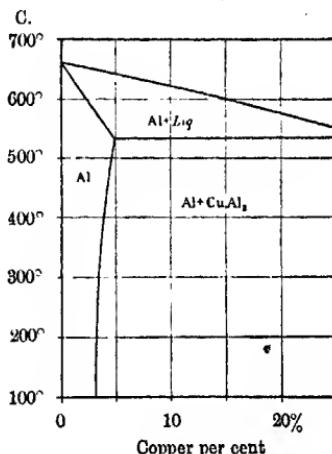
investigations, and attempts to solve these points have been included
in the present work.

The portion of the ternary system examined covers the range of
alloys containing up to 25 per cent copper and 45 per cent zinc.
Mixtures of three components can be represented by means of an
equilateral triangle, of which each corner represents one component.
All possible mixtures can then be represented by points within the
triangle. In Fig. 97 the shaded portion ABCD represents the part
of the ternary system which is here described. One hundred and
twenty alloys have been prepared and examined, and heating and

cooling curves have been taken on 63 of these. In the preparation
of these alloys, materials of very high purity have been used.

Experimental Methods.

The experimental methods and apparatus used are mainly those
which have for some time been adopted for researches of this kind
at the National Physical Laboratory. Some of these have already
been described ; others are described in connection with researches
on the constitution of light alloys both in the present Report and in
greater detail in papers dealing with this aspect of the work more
fully than is desirable here.

Constitution of the Alloys.

The results of the research may conveniently be described under
two main headings :—

1. The Completion of those portions of the binary systems
   necessary for the present research.
2. The Constitution of the Ternary Alloys.

Copper-Aluminium Alloys.

The copper aluminium alloys have been studied by a number of
workers ; the generally accepted diagram rests on the work of
Gwyer,[*] Carpenter and Edwards,[†] and of Curry,[‡] and may be
regarded as practically complete at the aluminium end. The
solubility of the compound $CuAl_2$ in solid aluminium at temperatures
below the solidus had not, however, been determined. This has
now been ascertained with some accuracy by a series of prolonged
annealing and quenching experiments for the temperature range
extending from the freezing point of the eutectic practically to room
temperature, and the modified and completed portion of the diagram so
obtained is given in Fig. 98 (page 199). It is found that approximately

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* Gwyer.   Zeit. Anorg. Chem., 1908, vol. lvii, page 113.
† Carpenter and Edwards.   Proceedings, I.Mech.E., 1907.
‡ Curry.   Journ. Phys. Chem., 1907, vol. xi, page 425.

5 per cent of copper is soluble in solid aluminium at 540° C., and that this amount decreases to about 3 per cent. at 20° C.

Zinc-Aluminium Alloys.

In this system, as established in the Tenth Report to the Alloys Research Committee, two points remained to be cleared up : (1) The cause of the small arrest on thermal curves at a temperature somewhat lower than the liquidus, and (2) the limits of existence of the β phase. The first point has been settled ; the small arrest has been found to be due to the presence of iron, which forms a eutectic with aluminium. This point may therefore be ignored in the study of the alloys under investigation. It is, unfortunately, impossible to use material which contains so little iron that this point is not found, since it appears when the iron content of the aluminium is less than $0\cdot1$ per cent.

With regard to the second point, a considerable amount of work has been devoted to an effort to clear up fully the constitution of the alloys in the region allotted to the "δ" phase in the existing diagram. Evidence has been obtained which shows clearly that under certain conditions the alloys consist of two phases in a part at least of the region between the horizontal lines at 443° C. and 256° C. The exact conditions under which this resolution into two phases takes place are not yet fully worked out, and publication of the results obtained is for that reason deferred.

The Copper-Zinc-Aluminium Alloys.

Within the limits of composition investigated in this research, no new ternary constituents have been detected in alloys in a state of thermal equilibrium ; in the case of cast alloys in the neighbourhood of the corner C, Fig. 97 (page 199), some constituents have been observed, probably belonging to the binary systems, which could not be identified with certainty without considerably extending the field of the research. Their presence in these alloys is due to inequalities in composition arising from micro-segregation during

solidification*; when the alloys are annealed so as to secure uniformity of composition, these constituents disappear.

Thermal curves have been obtained from a series of alloys containing 5, 10, 15, 20 and 25 per cent. copper, and for each of these concentrations of copper, zinc has been increased in steps of 5 per cent. from 0 to 45 per cent.

On the basis of the data so obtained, it has been possible to

FIG. 99.
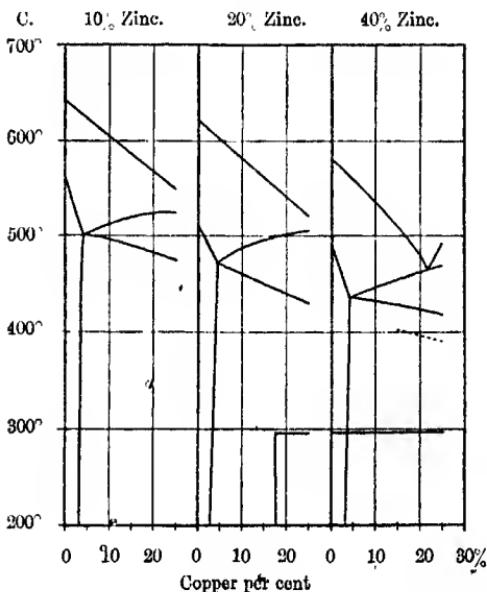
construct diagrams which represent sections through the ternary model parallel to the binary planes. Examples of such sections are shown in Figs. 99 and 100, the former representing sections parallel to the copper-aluminium plane, and the latter sections parallel to that of zinc-aluminium. Fig. 99 and other diagrams of sections parallel to it, thus indicate the equilibria of groups of alloys having a zinc-content which is constant for each section; in the case of Fig. 100 it is the copper-content which remains constant. In these

sections the full lines are those which rightly belong to the equilibrium diagram; the broken lines represent transformations which, although indicated by arrests on the cooling curves, have been found on further investigation to be due to metastable conditions which disappear on prolonged annealing.

Fig. 100.
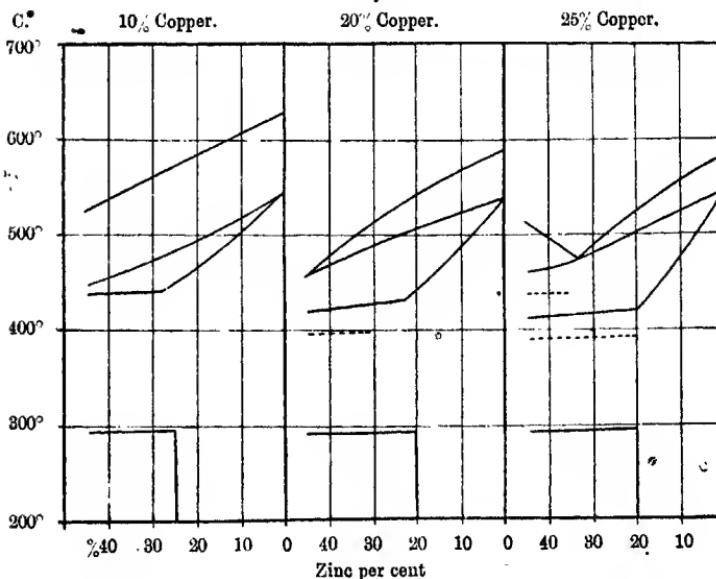
Constitutional Model.

From the observations made, and by the aid of parallel sections similar to those given in Figs. 99 and 100, a solid model may be constructed representing the constitution of the alloys at any temperature. In the case of a binary system of alloys the constitution may be represented on a plane diagram. Where three components exist two dimensions are needed to represent composition; a constitutional model may then be made on a triangular base, temperature being measured in the third dimension. A method

has already been described * whereby such a model may be constructed from different coloured wires in such a manner that the constitution under any conditions of composition and temperature can readily be seen. A photograph of such a model for this system of alloys is shown in Fig. 101, Plate 19. Since the constitution of the alloys can most readily be understood from a study of the model, it will be described by means of shaded drawings, the plane sections like Figs. 99 and 100 being retained so that, when necessary, they may be used for more accurate reference to the temperatures and concentrations which limit the transformations.

Liquidus Surface.

The general shape of the liquidus surface is shown in Fig. 102. The point A represents the melting point of pure aluminium and the lines AB, AE, the effect respectively of the addition of copper and zinc on the initial freezing point in these alloys. The beginning of the separation of the solid solution of aluminium containing copper and zinc is represented by the sloping surface ABCDE. The boundary line CD is the base of a valley beyond which the surface CDF, rising towards F, represents the separation from the liquid of $CuAl_2$. The projection of the liquidus surface on the base plane is shown in Fig. 105 (page 208). All alloys whose compositions lie within the area ABCDE commence to solidify by the deposition of crystals of aluminium solid solution. Those whose compositions lie within the area CDF commence to solidify by depositing crystals of $CuAl_2$.

The End of the Primary Separation.

The end of the primary separation is represented in Fig. 103 by the shaded surfaces AGLM and HGLDKC, which meet along the line GL. Alloys whose temperature ordinates cut the surface AGLM become completely solid at the temperature of that surface, and then consist, if in equilibrium, of a homogeneous solid solution of copper and zinc in aluminium. Alloys whose temperature ordinates

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* W. Rosenhain. Journal Institute of Metals, 1920, page 247.

Fig. 104.
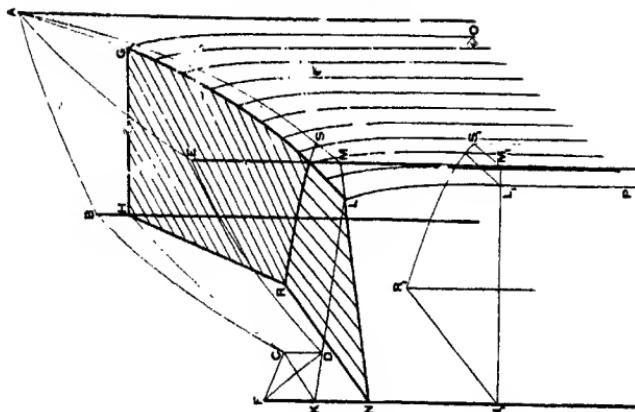
Fig. 103.
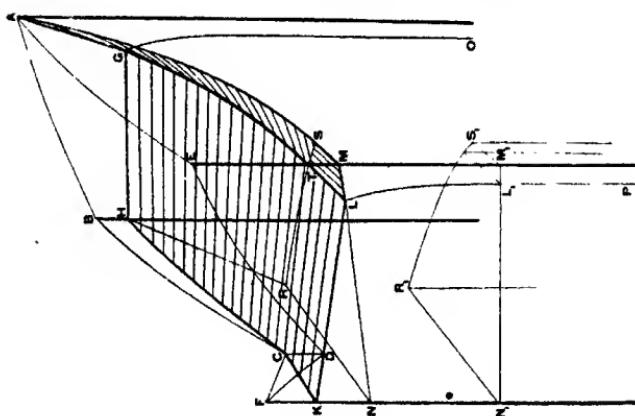
Fig. 102.
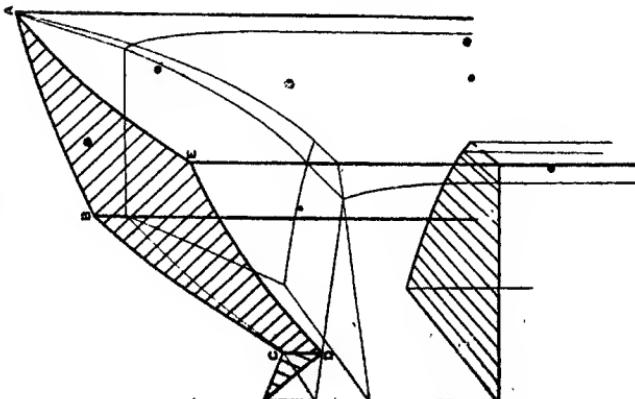
cut the surface HGLBC consist, at the end of the first period of solidification, of solid solution and liquid ; those whose temperature ordinates cut the surface DKC consist at the end of this period of CuAl$_2$ and liquid.

Second Stage of Solidification.

Alloys whose temperature ordinates cut the surface HGLDKC become solid in two stages. The second stage, which commences at this surface, consists in the separation of a binary complex * corresponding to the binary eutectic of the copper-aluminium series and consisting of the same phases (aluminium and CuAl$_2$) containing zinc in solid solution. Neither the composition of this complex, nor its temperature of solidification are constant, except along the line HG, where the system is one of two components only. Along the line GL, which represents the limit of solubility of CuAl$_2$ in the aluminium rich solid solution, the separation of CuAl$_2$ in infinitely small amounts takes place over an infinitely small temperature range, but the temperature of separation falls steadily as the concentration of zinc is increased. Elsewhere solidification during this stage takes place over a range of temperature ; this range is bounded by the surface HGLDKC and the broken surface HGTLNR (the latter shown shaded in Fig. 104) the edges HG and GL being common to the two surfaces.

Over the surface RSMLN, Fig. 104 the last stage of the solidification consists in a reaction between the solid solution which has already separated and the remaining liquid with the formation of the compound Zn$_3$Al$_2$. In the pure zinc-aluminium series this reaction is only found when the zinc content exceeds 40 per cent. As the amount of copper in the alloys is increased, however, the

* In a pure binary system of alloys a eutectic consists of two constituents which separate *at a constant temperature.* In the presence of a third element the same mixture of constituents separates over a range of temperature, and the word eutectic is not strictly applicable under such circumstances. The term binary "complex" has been introduced to describe such a mixture of constituents. The use of the expression ternary eutectic is strictly correct since this mixture of three constituents separates at a constant temperature.

reaction is found in alloys which possess a lower zinc content. This
is indicated by the slope of the boundary line RS towards the binary
copper-aluminium face of the model. The position of the surface
RSMLN is indicated by arrest points on the thermal curves. This
is the solidus of the alloys over this range of composition.

Changes in Solid Alloys.

Changes in the alloys after they have become solid may be
discussed under two heads :—

1. As the alloys containing more than 3 per cent. of copper
are cooled, separation of CuAl₂ takes place from the solid solution,
and the composition of this solid solution is indicated at any
temperature by the surface GOPL (Fig. 104), which divides the
alloys in which CuAl₂ occurs as a separate constituent from those
in which it is not so found.

2. The second change in the solid alloys occurs at the surface
R₁S₁M₁L₁N₁ (Figs. 103 and 104). At the points indicated by this
surface the decomposition of the compound Zn₃Al₂ takes place,
with the production of two solid solutions (the α and γ solid solutions
of the zinc-aluminium system). The temperature at which this
reaction takes place is but little affected by the addition of copper
to the system. It occurs at about 250° C. on cooling and 290° C.
on heating.

The shapes and extent of the various phase field boundaries are
indicated more clearly in the projections shown in Figs. 105, 106 and
107 (page 208). Fig. 105 shows the liquidus surfaces projected on to the
base of the model. The thick line DC marks the eutectic valley between
the aluminium and CuAl₂ surfaces ; the thin lines are temperature
contours (isothermals) on those surfaces. In the same way Fig. 106 is
a projection of the boundary between the first and second stages
of solidification. The surface BGLF shows the contours on the upper
surface of the eutectic field. The surface AGLE is the solidus of
the aluminium-rich solid solution. Fig. 107 shows the constitution
of the solid alloys at room temperature. Alloys whose compositions
fall within the area AGTS consist entirely of solid solution. Those

FIG. 105.—*Projection of Liquidus Surface.*
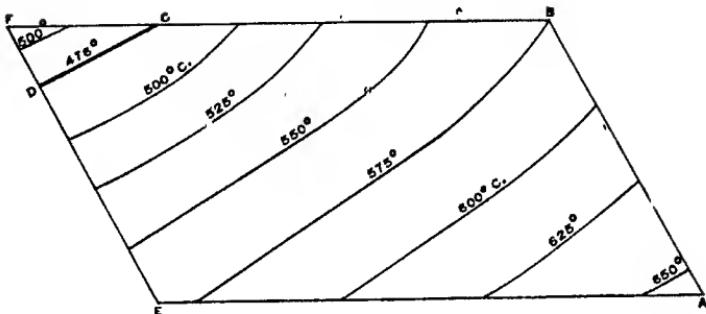
FIG. 106.—*Projection of End of Primary Separation.*
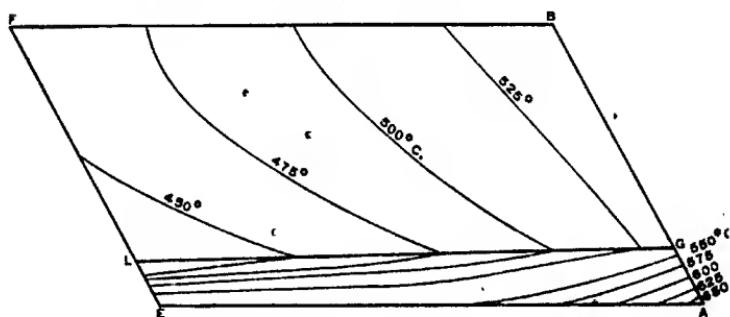
FIG. 107.—*Constitution of Solid Alloys at Room Temperature.*
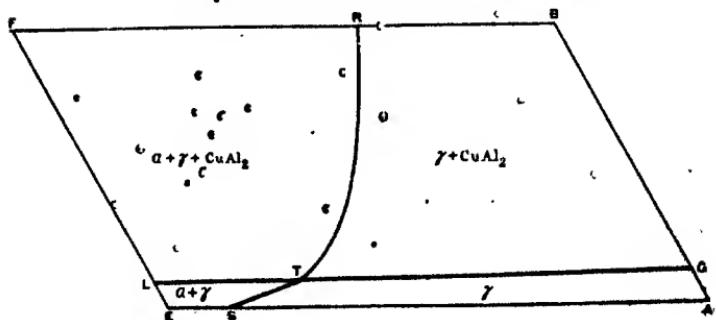
falling witnin the area GTRB contain aluminium-rich solid solution
and CuAl$_2$. Alloys falling within t' area STLE consist of aluminium-
rich solid solution and zinc-rich solid solution derived from the
decomposition of the aluminium-zinc compound. Alloys falling
within the area LTRF consist of the three constituents—aluminium
containing dissolved zinc and copper, zinc-rich solid solution as
above containing dissolved aluminium and copper, and CuAl$_2$.

Microstructures.

The microstructures of the alloys of this system are illustrated
in Figs. 108 to 119, Plates 16–19. Fig. 108, Plate 16, shows an alloy
containing 40 per cent of zinc and 10 per cent of copper, as cast. It
consists of primary crystals of aluminium containing copper and zinc
in solution, and CuAl$_2$. The primary crystals have a cored structure,
and some parts appear dark in the photograph. This dark appearance
is due partly to the decomposition of the solid solution at about
250° C. On annealing such an alloy the cores disappear, and it
consists entirely of aluminium-rich solid solution and CuAl$_2$. Fig.
109, Plate 16, shows an alloy containing 30 per cent of zinc and
20 per cent of copper as cast. The same constituents appear in
this alloy, but the amount of CuAl$_2$ is increased, as is also the
extent of decomposition of the solid solution. Fig. 110, Plate 16,
shows an alloy containing 30 per cent of zinc and 25 per cent
of copper. The composition of this alloy is represented by a
point close to the line DC in Fig. 105, and it therefore consists
almost entirely of a eutectic containing CuAl$_2$ (light etching) and
aluminium-rich solid solution which has decomposed subsequent to
its formation (dark etching). Fig. 111, Plate 16, shows an alloy
containing 45 per cent of zinc and 20 per cent of copper and is
very similar to Fig. 110, except that in this case separation of
CuAl$_2$ as a primary constituent has taken place.

The changes which take place in these alloys as a result of heat
treatment may be shown with reference to the alloy containing 30
per cent of zinc and 20 per cent of copper, which has been illustrated
in Fig. 109. Fig. 112, Plate 17, shows the same alloy under a higher
magnification : in this photograph the duplex structure of the dark

etching portions can be seen, while at the same time there occur, associated with these dark etching portions, small pieces of other constituents whose identity has not been determined. On re-heating such an alloy for a short, time above the critical -point at about 250° C. followed by quenching, the structure shown in Fig. 113, Plate 17, is obtained. In this alloy duplex portions of the solid solution have become homogeneous, but the small quantities of unidentified constituent referred to above are still present. Prolonged annealing, however, results in the disappearance of these constituents; Fig. 114, Plate 17, shows the structure of this alloy after annealing for twenty hours at 405° C. followed by quenching. After this treatment the alloy consists entirely of aluminium-rich solid solution and $CuAl_2$ (white). On raising the temperature till higher to 425° C. followed by quenching, the structure shown in Fig. 115, Plate 17, is obtained; melting of the alloy has commenced at the boundaries of the crystals. The temperature employed in this annealing experiment is therefore somewhat higher than the solidus of this alloy. On raising the temperature still further, the amount of liquid increases, as is shown in Fig. 116, Plate 18, which represents the alloy quenched from 440° C.

Fig. 117, Plate 18, shows the structure of the alloy containing 20 per cent of zinc and 10 per cent of copper after prolonged annealing at 420° C. followed by slow cooling to 220° C. from which temperature it was quenched. This alloy in the "as cast" condition shows considerable coring of the solid solution, and large dark-etching patches due to its decomposition. Long annealing produces uniformity of composition throughout the solid solution, and in this case decomposition does not occur in any portion after cooling through the transformation temperature. Fig. 118, Plate 18, represents the alloy containing 25 per cent of zinc and 10 per cent of copper after similar treatment. In this case, although the zinc content is only 5 per cent higher, decomposition of the solid solution has taken place throughout the whole mass of the material. It is evident then, that these two alloys lie one on either side of the line RT, Fig. 107 (page 208). Fig. 119, Plate 19, shows the alloy containing 35 per cent of zinc and 20 per cent of copper after the same treatment.

Fig. 120.—Iron-Aluminium.

Iron per cent
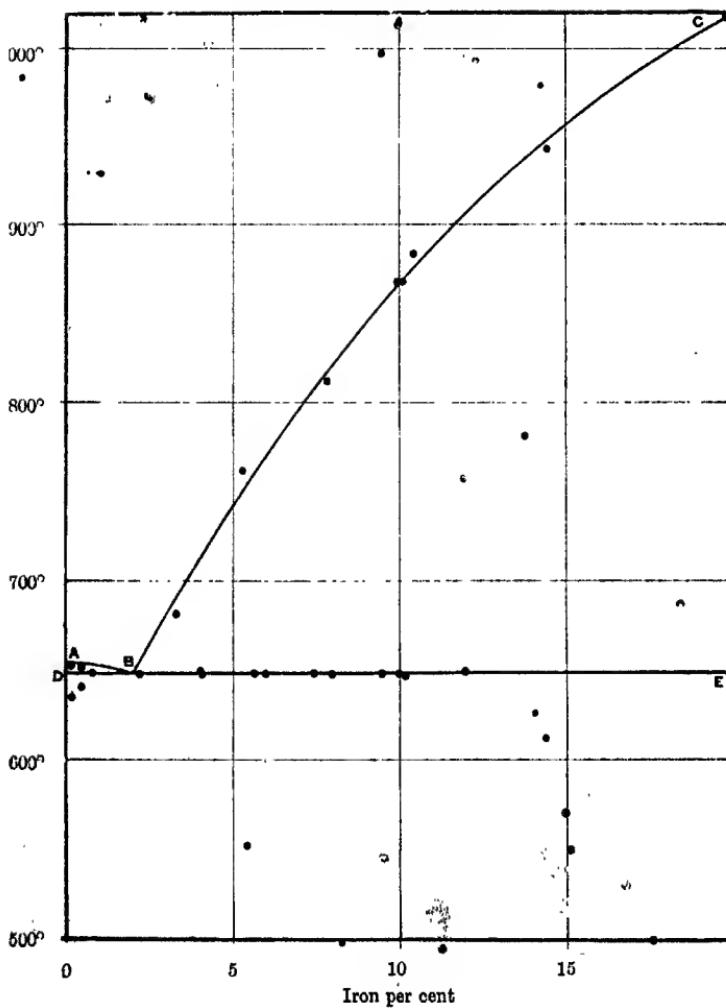
The duplex character of the decomposed solid solution is very clear in this case.

V(b). THE CONSTITUTION OF THE SILICON-IRON-ALUMINIUM ALLOYS.

The Iron-Aluminium Alloys.

It appeared necessary as a preliminary to the investigation of the ternary system to review the binary system, iron-aluminium, at the aluminium end.   This has previously been investigated over the whole range of composition by Gwyer, who found that iron forms a compound $FeAl_3$ with aluminium; his diagram is, however, not complete at the aluminium end.

For the present purpose alloys have been prepared in which the iron content varies from 0-20 per cent, and cooling curves have been determined on these alloys.   The results are expressed in the diagram shown in Fig. 120 (page 211).   The compound $FeAl_3$ forms with aluminium a eutectic containing 2 per cent of iron and melting at 648° C. This eutectic temperature is very close to the melting point of pure aluminium, and it is practically impossible, in the case of alloys containing less than the eutectic percentage of iron, to distinguish between the arrest due to the separation of primary aluminium and that due to the separation of the eutectic.   With alloys of higher iron content, the liquidus arrest point rises rapidly, and in the case of an alloy containing 20 per cent of iron, separation of $FeAl_3$ begins at a temperature higher than 1,000° C. $FeAl_3$ is practically insoluble in solid aluminium, and no sample of this metal which the Authors have examined is quite free from particles of the free compound.

Silicon-Iron-Aluminium Alloys.

In order to investigate the constitution of this system, five different series of alloys containing 1 to 8 per cent iron have been prepared, the iron content being maintained at a fixed value in each series, while silicon was present in proportions up to 8 per cent.

Each of these series of alloys lies, therefore, in a plane of the ternary model of the system parallel to the aluminium-silicon face. Thermal curves have been taken of these alloys, and the arrest points obtained are shown in the diagrams in Figs. 121 to 125. These diagrams represent sections of the ternary model which have been arrived at partly by the aid of thermal curves and partly as a
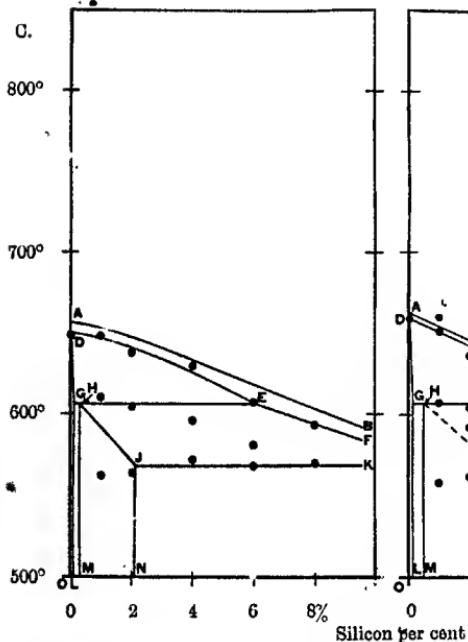
FIG. 121.—1% *Iron.*   FIG. 122.—2% *Iron.*
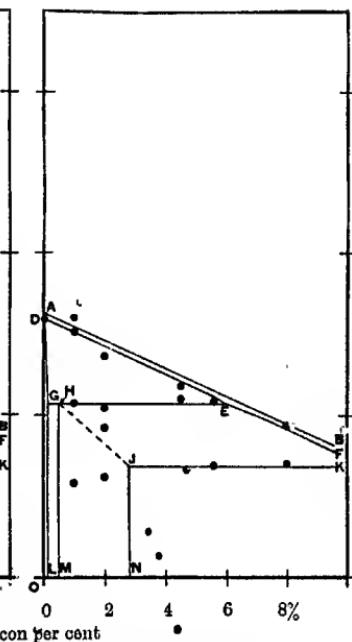
result of the microscopic examination of heat-treated specimens. Fig. 121 represents the section corresponding to a constant iron content of 1 per cent. The first separation from the liquid takes place along the line AB and consists of aluminium-rich solid solution. The point A in this section corresponds to a point on the line AB in Fig. 120. The line DE indicates the separation of the aluminium iron binary complex. At the temperature of the line GE, the

compound FeAl₃, which has already separated as part of the binary complex, reacts with a portion of the liquid to form a new constituent, which will be referred to as constituent X. At a lower temperature, namely, that of the line JK, a ternary eutectic solidifies, consisting of the constituent X, aluminium and silicon. The line EF appears to represent the separation of a binary complex consisting of X and aluminium. The next section, Fig. 122 (page 213), represents alloys containing 2 per cent of iron. This is very similar to the previous section. Fig. 123 (page 215), is the section representing alloys containing 4 per cent of iron. In this series the primary separation no longer consists of aluminium-rich solid solution; the point P belongs to that part of the aluminium-iron binary diagram where FeAl₃ separates from solution, and this same constituent appears to separate along the line PQ. At the point Q, however, the direction of the liquidus changes, as is shown by the dotted line QR. It seems probable, therefore, that over this portion a different constituent separates, but the Authors have not as yet carried their investigation sufficiently far to identify this new constituent. In the sections shown in Figs. 124 and 125 (page 215), the same characteristics appear. The actual changes which take place in the alloys between the line PQR and line DEF have not been fully investigated. Up to the point Q they appear to be simple, and between this line and the line DE, separation of FeAl₃ appears to take place continuously. Below the line QR, however, other reactions occur, giving rise to arrest points on the cooling curves. These are indicated by plotted points in the sections representing alloys containing 6 per cent of iron (Fig. 124), and 8 per cent of iron (Fig. 125). The constituents involved all have a very similar appearance under the microscope, and are very difficult to distinguish from FeAl₃. In view of the fact that this portion of the diagram does not represent alloys of practical importance, and since the elucidation of the constitution over this range of composition appeared likely to take considerable time, this work was suspended in order that more urgent problems might be investigated. In that part of all sections representing alloys containing up to about 4 per cent of silicon, the constitution is believed to have been satisfactorily determined, and is indicated

Fig. 125.—8% Iron.
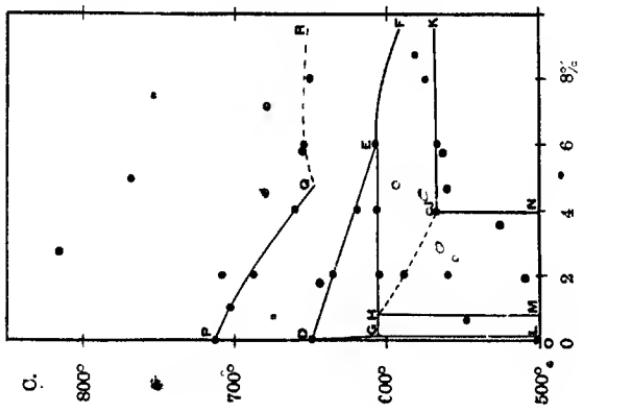
Fig. 124.—6% Iron.

Fig. 123.—4% Iron.

Silicon per cent
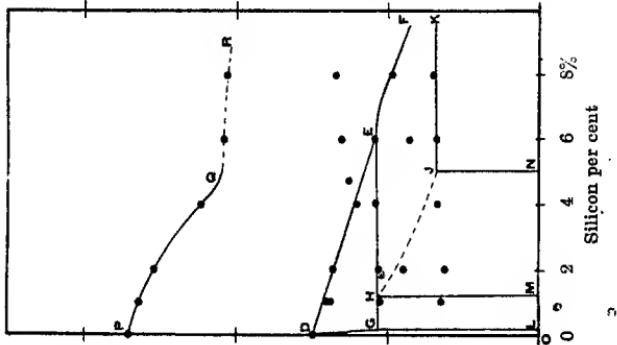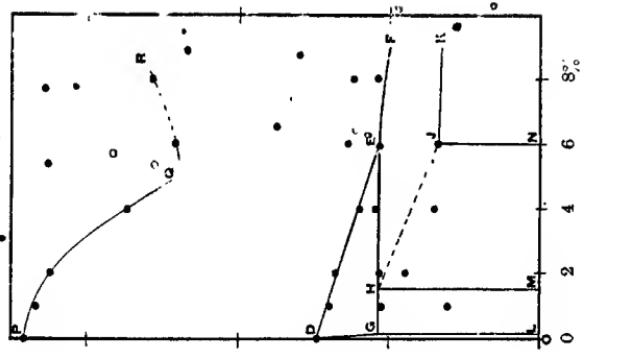
by the different phase-fields shown in the sections in Figs. 121 to 125.

The small area DOLG on each section represents alloys consisting of aluminium and FeAl$_3$. This is an extremely narrow section running parallel to the aluminium iron binary face of the constitutional model, so narrow in fact that it is not possible accurately to determine its position on account of the presence of a certain amount of silicon as an impurity in all the alloys used. The area LGHM represents alloys in which aluminium rich solid solution, FeAl$_3$ and the constituent X are all present, while the area MHJN represents alloys in which all the FeAl$_3$ which first separates has been converted into the constituent X. Figs. 126 and 127, Plate 20, represent the microstructure of alloys in each of these sections respectively. Fig. 126 shows an alloy containing 1 per cent of silicon and 8 per cent of iron (magnification × 1,000); the alloy consists of a matrix of aluminium rich solid solution containing a long crystal of FeAl$_3$ surrounded by an envelope of the constituent X. Fig. 127 which represents the alloy containing 2 per cent of silicon and 8 per cent of iron (× 1,000), shows aluminium rich solid solution and the constituent X. These alloys were photographed after a prolonged period of annealing (170 hours) at 565° C.

In the area lying beyond the lines JK and JN, silicon occurs as a separate constituent. It separates as one of the constituents of the ternary eutectic at a temperature of 570° C. This ternary eutectic has not a typical " eutectic " appearance, since it consists largely of aluminium, and contains but small quantities of silicon and the constituent X. Fig. 128, Plate 20, shows an alloy containing 3 per cent silicon and 1 per cent iron (magnification × 1,000). Silicon appears as the darkest constituent.

It follows from the results of this investigation that most commercial aluminium (and its alloys) do not contain, as has usually been supposed, the free compound FeAl$_3$, except possibly as a metastable constituent. The proportion of silicon present is usually sufficient to convert all the FeAl$_3$ which first separates into the constituent X.

Fig. 29.—*Aluminium-Magnesium Alloys.*
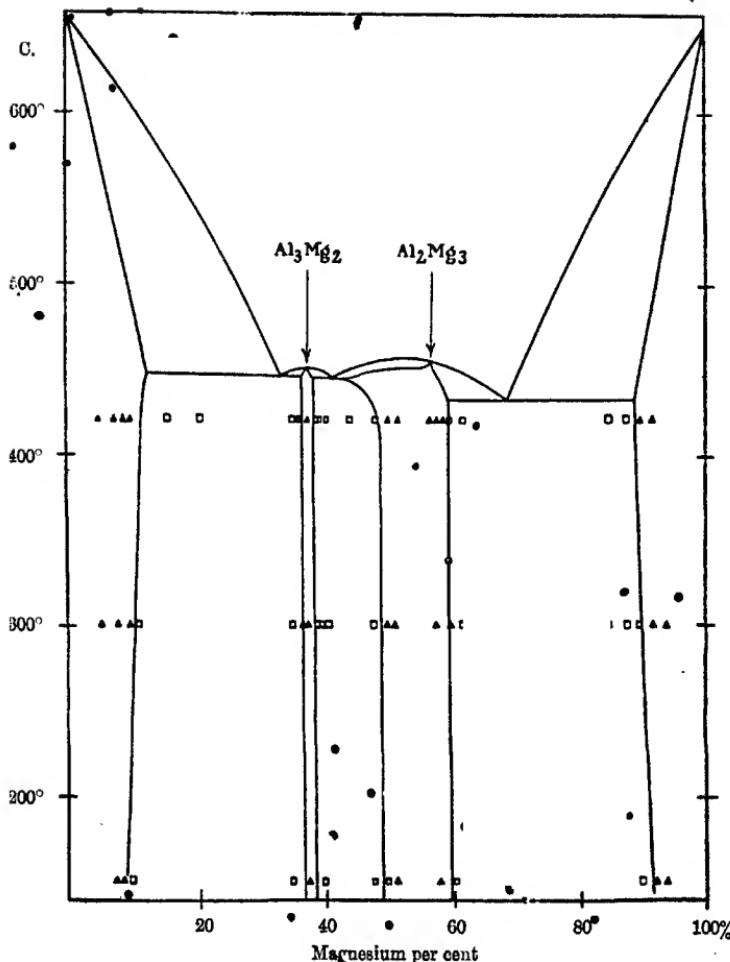
$V(c)$. The Alloys of Aluminium with Magnesium, with Silicon,
AND WITH MAGNESIUM AND SILICON.

Aluminium-Magnesium Alloys.

The only constitutional diagram of the aluminium-magnesium
alloys previously available was due to Grube[*] and was admittedly

Fig. 130.—*Aluminium-Magnesium Alloys.*

▲ INDICATES HOMOGENEOUS ALLOY

☐   „   DUPLEX ALLOY
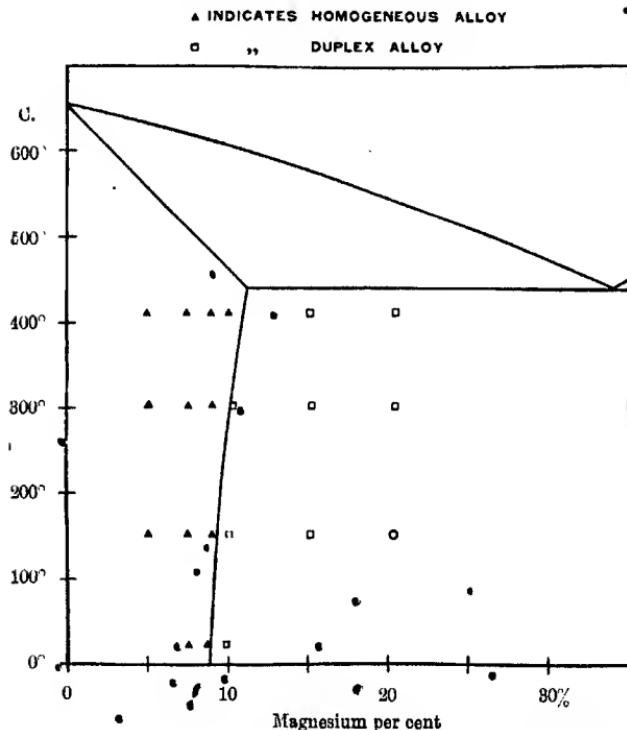
ncomplete since it was based on the results of thermal curves alone.
While for the particular subject under discussion an investigation
nto the whole series of alloys was not necessary, yet it was thought

* Grube.   Zeit. Anorg. Chem., 1905, vol. xlv, page 225.

in view of future possible extensions in the application of magnesium
alloys that it would be desirable to redetermine the constitution
over the whole system. This has therefore been done, and the new
diagram is shown in Fig. 129 (page 217). This is based not only on the
result of a large number of thermal curves, but also on the microscopic
examination of a large number of specimens after special prolonged
heat treatment. The full details of this treatment have been
separately described * and reference should be made to this Paper
for the more complete account of the methods used. The main
differences between this diagram and that of Gruhe are found over
the composition range extending from 33 to 45 per cent magnesium.
In this region it is shown that a compound Al_3Mg_2 exists, which
forms a eutectiferous system both with aluminium and with Al_2Mg_3.
Near the ends of the system, the solubility of Al_3Mg_2 in aluminium
and Al_2Mg_3 in magnesium have been accurately determined over a
wide range of temperature. It is found that solid aluminium will
hold in solution approximately 11 per cent of magnesium at a
temperature of about 420° C. At lower temperatures the solubility
of this compound decreases slightly and is about 9·5 per cent
at 150° C. Fig. 130 shows the data on which the solubility curve is
based. Up to these percentages, therefore, aluminium-magnesium
alloys consist of homogeneous solid solution, though it by no means
follows that the alloys in the cast state have this constitution. Fig. 131,
Plate 21, represents an alloy containing 10 per cent magnesium.
This alloy has been annealed at 420° C., after which treatment it
was homogeneous ; it has subsequently been slowly cooled from
420° C. to 160° C., during which treatment some particles of Al_3Mg_2
(dark etching in the photograph) have separated from solution.
Fig. 132, Plate 21, shows the structure of an alloy containing
20 per cent magnesium. Fig. 133, Plate 21, represents an alloy
containing 35 per cent magnesium. In this case it consists largely of
$Al_3 Mg_2$ (dark etching), while the amount of solid solution is small.
This alloy consists of practically pure eutectic.

In the case of the alloys rich in magnesium, the compound

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* Hanson and Gayler. Journal Institute of Metals, 1920, vol. ii.

Al_2Mg_3 does not appear as a separate constituent at 420° C. until about 10 per cent aluminium is present, this solubility decreasing slightly as the temperature is lowered to about 9 per cent at 150° C. In this case also, chill cast samples contain free Al_2Mg_3 when the aluminium content exceeds about 5 per cent. A photograph of

FIG. 135.—*Aluminium-Magnesium Alloys.*
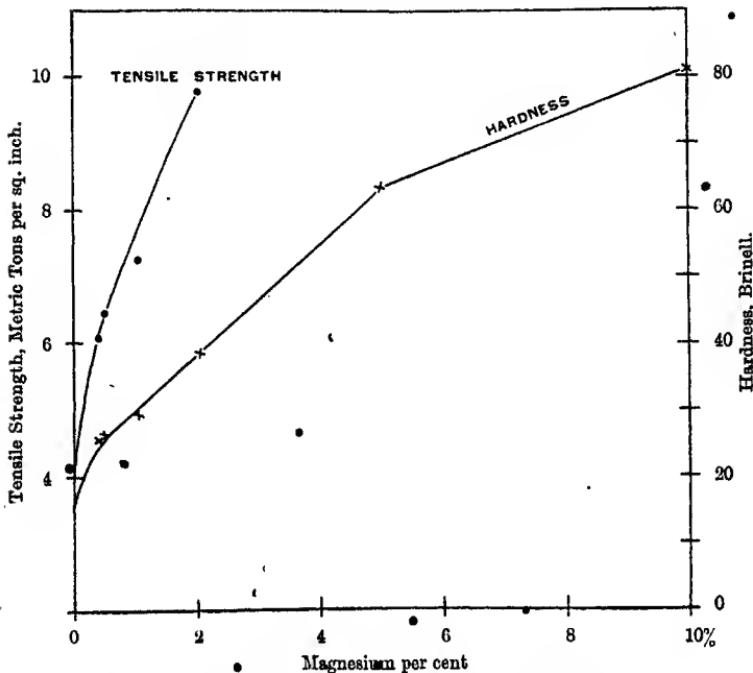
the 10 per cent alloy chill cast is shown in Fig. 134, Plate 21. On annealing, this sample becomes homogeneous.

Mechanical Properties of Aluminium-Magnesium Alloys.

A number of mechanical tests have been carried out to determine the effect of magnesium on aluminium; the results are plotted in Fig. 135. It is seen that magnesium has a pronounced effect in

increasing the mechanical strength of aluminium. For example, the addition of 2 per cent raises the hardness, as measured by the Brinell method, from 15 to 39, while the tensile strength is increased from 4 tons to nearly 10 tons per square inch. With higher percentages the alloys become still harder, and in fact they are

FIG. 136.—*Aluminium-Silicon Alloys.*
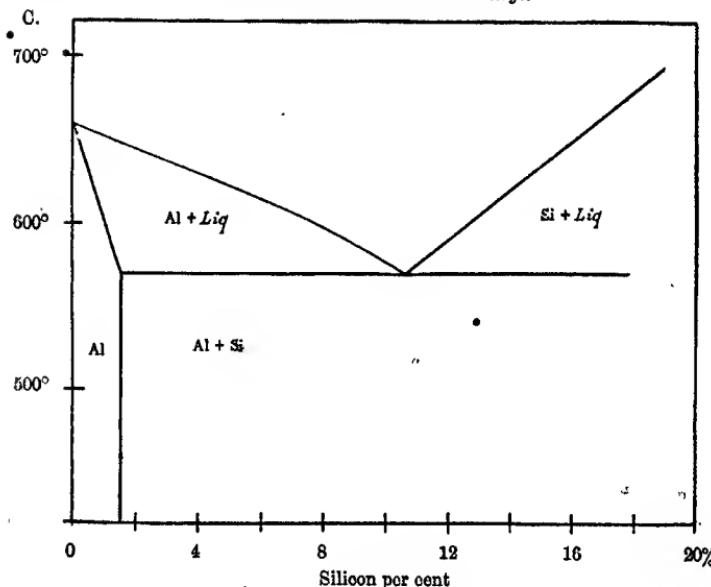
extremely difficult to forge or roll when more than 4 or 5 per cent of magnesium is present.

Aluminium-Silicon Alloys.

Aluminium and silicon form a simple eutectiferous series with a eutectic point at about 10·5 per cent silicon. Silicon itself is sparingly soluble in solid aluminium, the maximum amount which it has been possible to bring into solution being about 1·5 per cent; this solubility does not appear to decrease appreciably as the temperature is lowered. The equilibrium diagram is shown in Fig. 136. A photomicrograph of an aluminium-silicon alloy containing

3 per cent silicon is shown in Fig. 137, Plate 21, in which the silicon appears as the darker constituent. This sample is unetched, since the bluish-grey colour of the silicon makes it possible to distinguish it from aluminium without special etching of the specimen. An interesting point has been observed in connection with the chemical behaviour of aluminium alloys containing silicon. When these

FIG. 138.—*Aluminium-Silicon Alloys.*
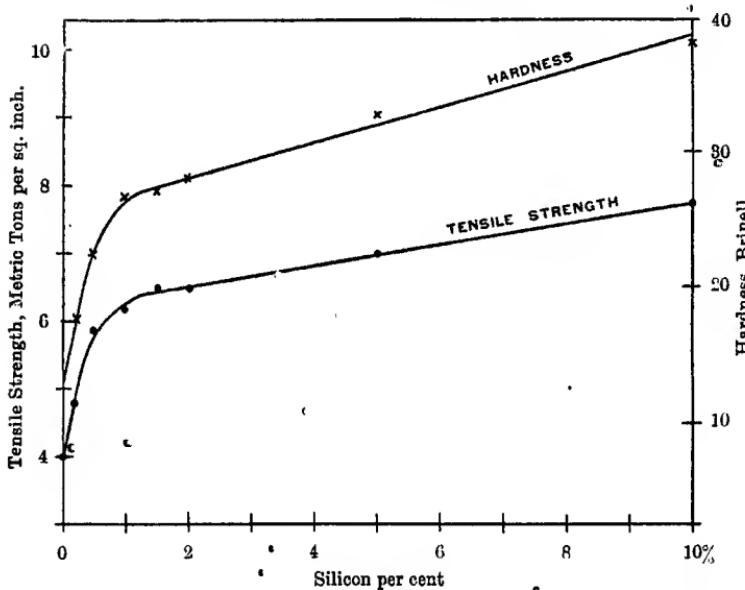
are attacked with strong oxidizing agents in the ordinary methods of chemical analysis, it frequently happens that part of the silicon remains unattacked and is returned as free silicon, while part is oxidized and is returned as silica. In these alloys it has been shown that the portion which is oxidized to SiO_2 is that which existed in the state of solid solution in the alloy (or in a state of combination), while that portion which is returned as free silicon existed as crystals of free silicon in the alloy itself.
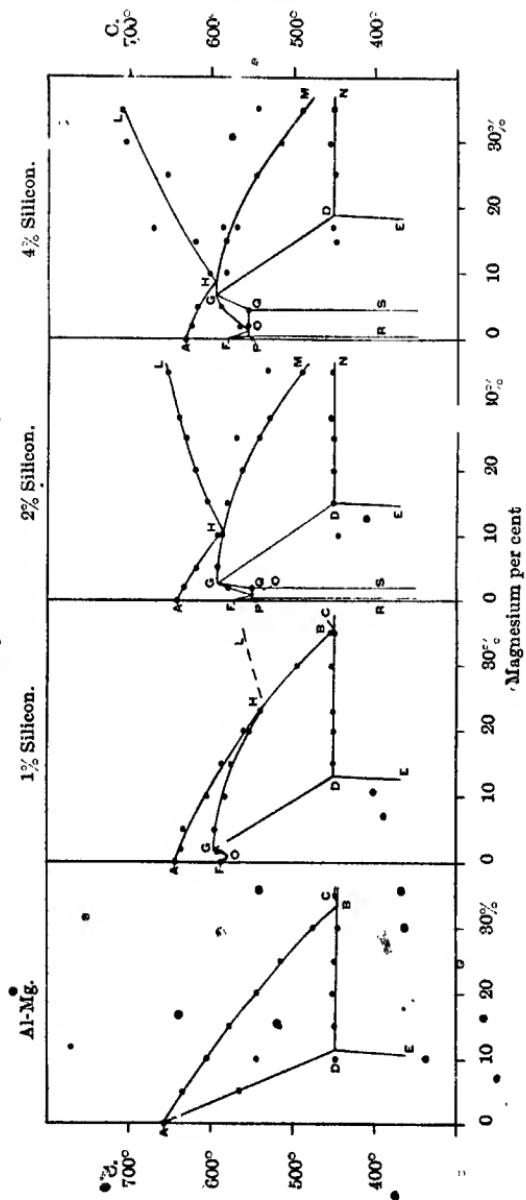
FIG. 139.—Silicon-Magnesium-Aluminium Alloys.

Mechanical Properties of Silicon-Aluminium Alloys

The effect of silicon on the mechanical properties of aluminium has been investigated over a fairly wide range of composition. Silicon-aluminium alloys are very readily forged and rolled, and these operations have been carried out on an alloy containing as much as 20 per cent silicon. Hardness tests on these alloys as forged and tensile tests on the alloys in the form of rolled strip $\frac{1}{16}$ in. thick are shown in Fig. 138 (page 222). The first effect of the addition of silicon is to increase the tensile strength and hardness very appreciably. This increase, however, becomes less rapid when the silicon content exceeds $1-1\frac{1}{2}$ per cent, a composition which corresponds with the limit of solubility of silicon in aluminium.

ALLOYS OF ALUMINIUM WITH MAGNESIUM AND SILICON.

Constitution.

The constitution of these alloys has been determined by means of thermal curves, and by the microscopic examination of heat-treated specimens. Thermal curves have been taken on alloys containing from 0 to 35 per cent magnesium and 0 to 11 per cent silicon, a total of about seventy alloys having been employed for this purpose. The results of these thermal curves are shown in the diagrams in Figs. 139 (page 223), and 140 (page 225). These diagrams represent sections through a ternary constitutional model, the sections being parallel to the aluminium-magnesium face ; they therefore represent a series of alloys in each of which the silicon content is kept constant, while the magnesium content is increased from 0 to 35 per cent. The modifications of the original binary diagram which result from the addition of silicon can be followed step by step, but they become somewhat complicated after the addition of more than 1 per cent of silicon, and are hardly intelligible until they have been collected together to form a ternary equilibrium model. It may, however, at this stage be mentioned that the points marked on the diagrams are actual observed points on thermal curves. These in themselves are not sufficient completely to determine the constitution of the

Fig. 140.—*Silicon-Magnesium-Aluminium Alloys.*
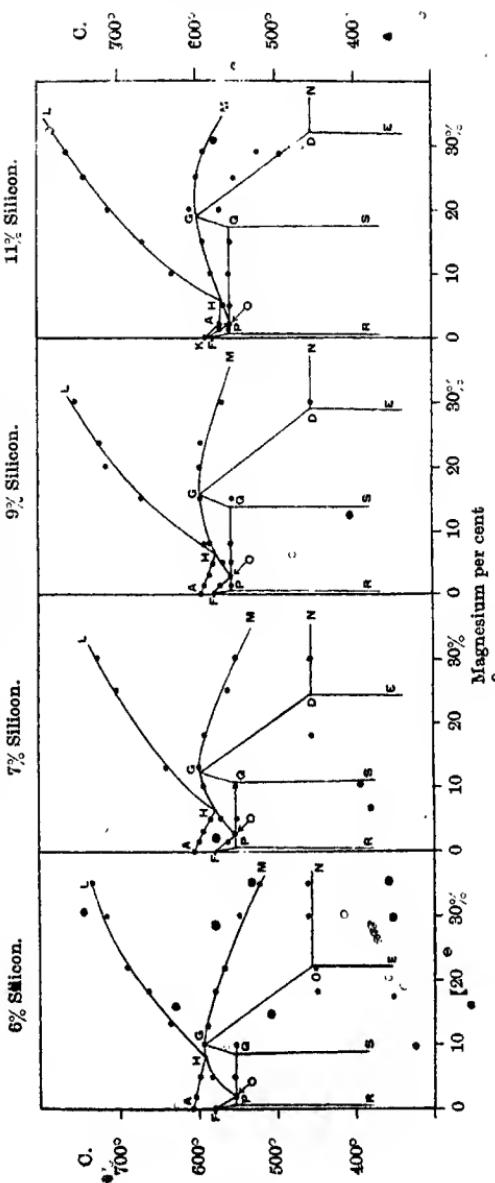
alloys since equilibrium conditions are not, as a rule, obtained with
rates of cooling suitable for the taking of thermal curves. In some
respects, therefore, chiefly with regard to the solubility of the
constituents in one another in the solid state, the diagrams have
been completed as a result of prolonged annealing of selected alloys
followed by microscopic examination. For example, the line ND,

Fig. 142.—*Silicon-Magnesium-Aluminium Alloys.*

Liquidus Surface.
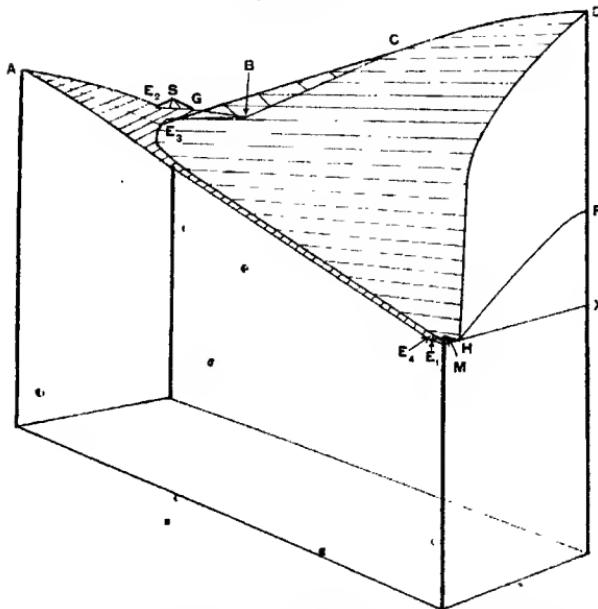
which appears in all the sections shown, would be prolonged beyond
the point D if cooling curve observations were relied upon to give
the final diagram.

The data obtained from these sections have been combined
together to form an equilibrium model, which represents
simultaneously the effect of variations of composition and
temperature on the constitution of the alloys. A photograph of

such a model, constructed in the manner which has already been referred to, is shown in Fig. 14b, Plate 22. This, however, is somewhat complicated and the photograph does not lend itself readily to detailed explanation, although the model itself is well suited for that purpose. A number of diagrams in perspective have therefore been prepared in which the various surfaces and phase

FIG. 143.—*Silicon-Magnesium-Aluminium Alloys.*

Second Stage of Solidification.
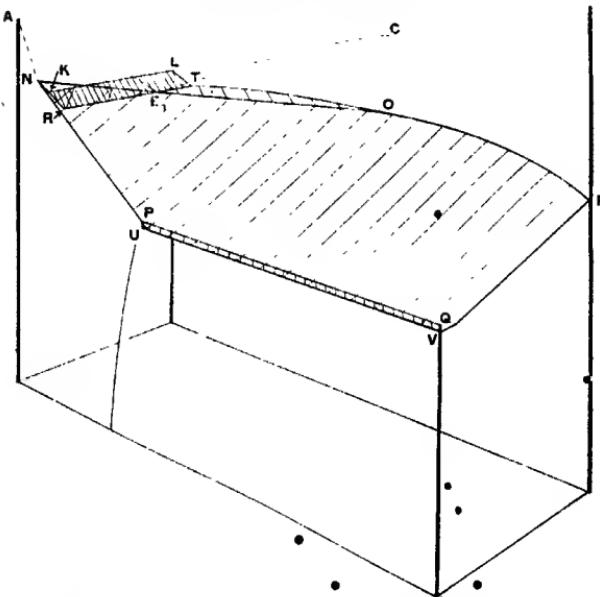
boundaries have been specially indicated by different methods of shading. Fig. 142, which shows one of these diagrams, represents the liquidus surfaces of the ternary system over the range of composition under investigation. This consists of four different surfaces, of which two, E_2GS and E_1HM, representing respectively the surfaces at which silicon and Al_3Mg_2 begin to separate, are very small. The remaining surfaces are $AE_2GBE_3E_4E_1$ and $DCBE_3E_4H$

which intersect along the line BE_3E_4 and form a valley which has a maximum point at E_3. The first of these surfaces represents the beginning of the separation of aluminium containing dissolved silicon and magnesium, while the other surface represents the points of first separation of a compound of silicon and magnesium having the formula Mg_2Si. The common line of intersection BE_3E_4 represents points at which a eutectic-complex of these two constituents begins to separate. These surfaces may be said to represent the first stages in the solidification of these alloys.

The second stage of solidification is shown in Fig. 143 (page 227). This shows three surfaces representing the separation of three different eutectic complexes. The surface KLTR slopes downwards from the binary eutectic line KL, and represents the beginning of the separation of the aluminium-silicon eutectic complex. The surface PQVU sloping downwards from the binary eutectic line PQ represents the separation of the aluminium-Al_3Mg_2 eutectic complex. The surface RTOFVUN is of more complicated shape. It slopes downwards on either side of the horizontal line NO, which lies wholly within the surface. On the one side it meets the aluminium-silicon complex surface in the line RT, while on the other side it meets the aluminium-Al_3Mg_2 complex in the line UV. This ridge-shaped surface represents the upper limit of the separation of a eutectic complex of aluminium and the compound Mg_2Si; its intersections with the other two eutectic surfaces, the lines RT and UV, are horizontal and lie in the two ternary eutectic planes which the system possesses, and which are shown in the next Figure.

Fig. 144 represents the end of the solidification (or solidus) of the alloys. This consists of seven distinct surfaces and completely covers the whole range of compositions in the alloy system. The surface KLZR represents the end of the separation of the binary eutectic (or complex) of aluminium and silicon, and alloys over this range of composition are completely solid after the end of the separation of this complex. The horizontal plane RZW represents the separation of a ternary eutectic of aluminium, silicon, and Mg_2Si, the two surfaces NOWR and NOYU represent the solidus of the aluminium-Mg_2Si eutectic complex, and have the same ridge-

shaped formation as has the upper limit of the separation of this
body. It may be noted at this stage that the line NO is common
to both the upper and lower limits of the separation of the aluminium-
Mg₂Si complex; in other words, along this line this complex is
strictly a eutectic, separating at a constant temperature and not
over a range of temperature. The points G shown in the different

Fig. 144.—*Silicon-Magnesium-Aluminium Alloys.*

Solidus of Alloys.
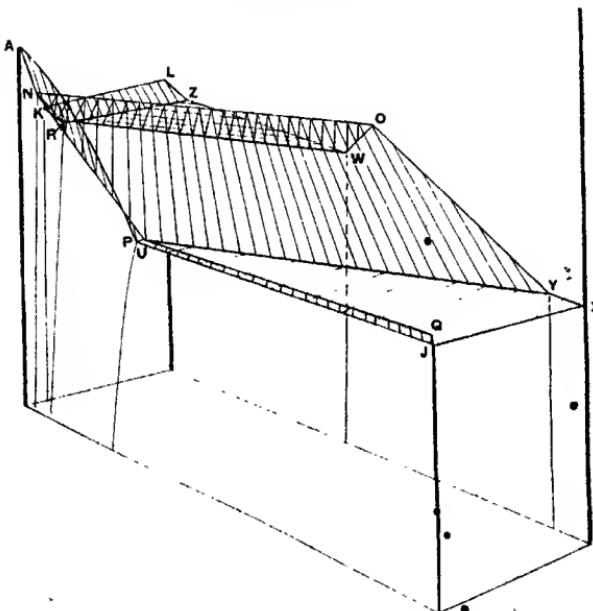
diagrams in Figs. 139 and 140 are points on this line NO. This fact will
be commented upon later. The surface NOWR meets the aluminium-
silicon-Mg₂Si ternary eutectic plane in RW. The surface NOYU
meets another ternary eutectic plane in UY. This latter ternary
plane UYXJ represents the separation of a ternary eutectic
consisting of aluminium, Al₃Mg₂ and Mg₂Si. The surface PQJU
represents the end of the separation of the binary eutectic complex

Fig. 145.—*Silicon-Magnesium-
Aluminium Alloys.*
Solidus of Alloys.
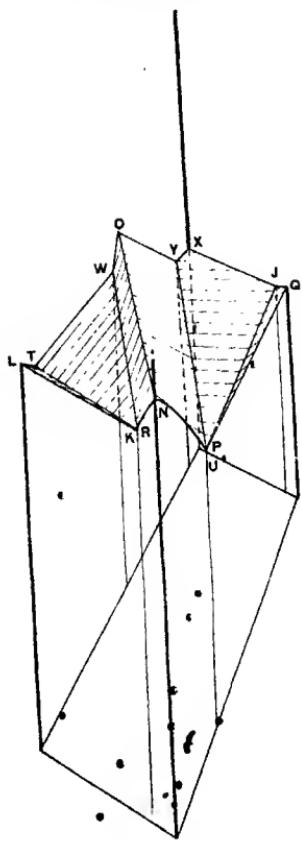
Fig. 146.—*Projection of Liquidus Surface.*
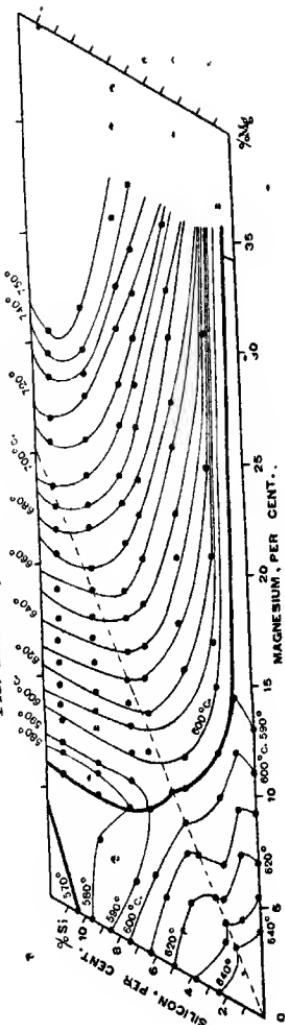
of aluminium and Al_3Mg_2 and intersects the corresponding ternary eutectic plane in UJ. Finally, the complicated surface AKRNUP, which slopes somewhat steeply from the point A, represents the solidus of the aluminium-rich solid solution containing dissolved silicon and magnesium.

Fig. 145 is another diagram showing the solidus surface when viewed from near the aluminium corner. This Figure illustrates particularly well the ridge-like shape of the solidus of the aluminium-Mg_2Si eutectic complex of which the horizontal line NO forms the crest. The solidus of the aluminium-rich solid solution is not drawn in this Figure.

Figs. 146, 147 and 148 show projections of the different surfaces of separation on the horizontal base of the model. Fig. 146 is a projection of the liquidus surfaces and the different areas show the ranges of composition over which the four constituents—aluminium, Mg_2Si, silicon, and Al_3Mg_2—are the first to separate from the molten metal. The thin lines are temperature isothermals. Fig. 147 (page 232), shows in like manner the ranges of composition over which the different binary eutectic complexes of the system separate, while Fig. 148 (page 232), is a projection of the constitution of the solid alloys immediately after the solidus has been passed.

Typical microstructures of these alloys are shown in Figs. 149 to 153, Plates 22 to 24. Fig. 149, Plate 22, shows an alloy containing 5 per cent magnesium and 4 per cent silicon. It consists of primary crystals of aluminium containing dissolved magnesium and silicon, a coarse binary eutectic complex consisting of aluminium and Mg_2Si, and a fine-grained ternary eutectic consisting of aluminium, Mg_2Si, and silicon. Reference to the projections of the solid model will show that this is the constitution which is indicated for an alloy of this composition. Fig. 150, Plate 23, shows an alloy containing 10 per cent magnesium and 4 per cent of silicon. In this section there is very little primary separation of aluminium, and the alloy consists very largely of a binary eutectic complex of Mg_2Si and aluminium. The composition of this alloy is very nearly that of the pure binary eutectic of these two materials ; a slight excess of aluminium is, however, to be observed. Fig. 151, Plate 23, shows the microstructure
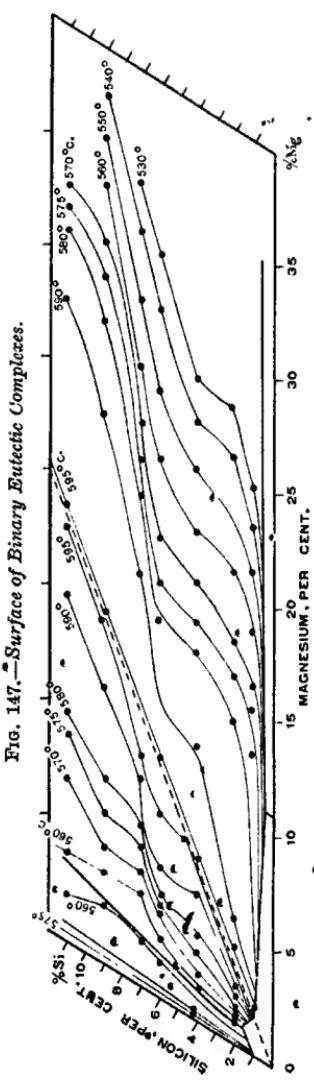
FIG. 147.—*Surface of Binary Eutectic Complexes.*
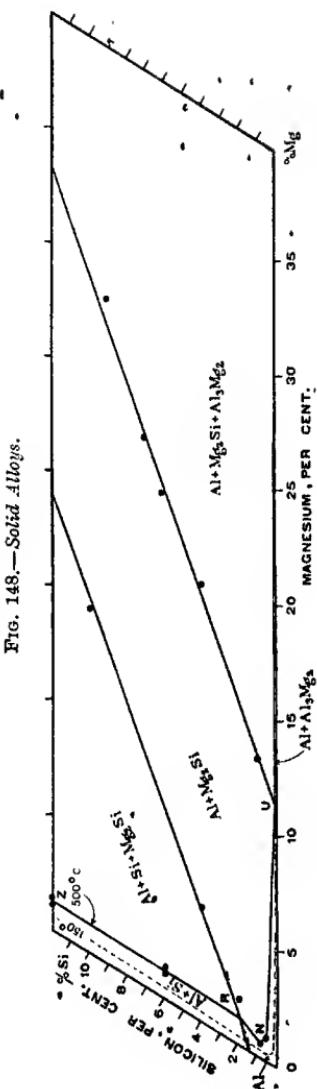
FIG. 148.—*Solid Alloys.*

of an alloy containing 10 per cent of magnesium and 3 per cent of silicon. This consists of a few large primary crystals of the compound Mg_2Si, a considerable proportion of eutectic complex consisting of Mg_2Si and aluminium and a small proportion of ternary eutectic of aluminium, Mg_2Si and silicon, which has an indistinct fine-grained structure. Fig. 152, Plate 23, shows an alloy consisting of 25 per cent magnesium and 2 per cent of silicon. In this alloy the primary separation is of crystals of Mg_2Si, which are shown as the dark etching constituent in the photograph. The remainder consists of binary eutectic complex of aluminium and Mg_2Si, and a ternary eutectic of Mg_2Si, aluminium and Al_3Mg_2, the latter constituent having a medium tone between the light aluminium and the dark Mg_2Si. The characteristic appearance of the binary eutectic is not always visible in alloys in which the separation is really that of a complex taking place over a wide range of temperature. Under these circumstances, deposition of the one constituent may take place on the primary crystals which have already separated during the first stage of solidification, and in consequence, the typical appearance which one usually associates with a eutectic may be lost.

Crystals of Mg_2Si in these alloys have a very beautiful and characteristic appearance, especially when they occur in large quantities as a primary separation. They have a beautiful blue colour, and appear iridescent in the polished specimen, in which they are visible quite readily without etching. Fig. 153, Plate 24, shows the appearance of an unetched alloy containing 35 per cent magnesium and 10 per cent of silicon. The dark crystals are Mg_2Si. The compound Al_3Mg_2, which also exists in this alloy, does not appear until the sample is etched.

A careful examination of the constitutional model reveals the fact that aluminium and Mg_2Si form with one another a system which has the properties usually associated with an ordinary binary system of alloys. For example, the two constituents form with one another a true eutectic along the line NO, which has a constant temperature of separation, and the typical appearance under the microscope of an ordinary binary eutectic. Elsewhere in the

system, the separation of this body takes place over a range of temperature, while the beginnings of its separation always occur at a lower temperature than that of the line NO. It would seem, therefore, that these two constituents form a simple binary system, whose base-line on the equilateral triangle representing the ternary system runs from the aluminium corner to the point of the Mg-Si line representing the compound Mg₂Si. The whole

FIG. 154.—*Aluminium-Magnesium-Silicide Alloys.*
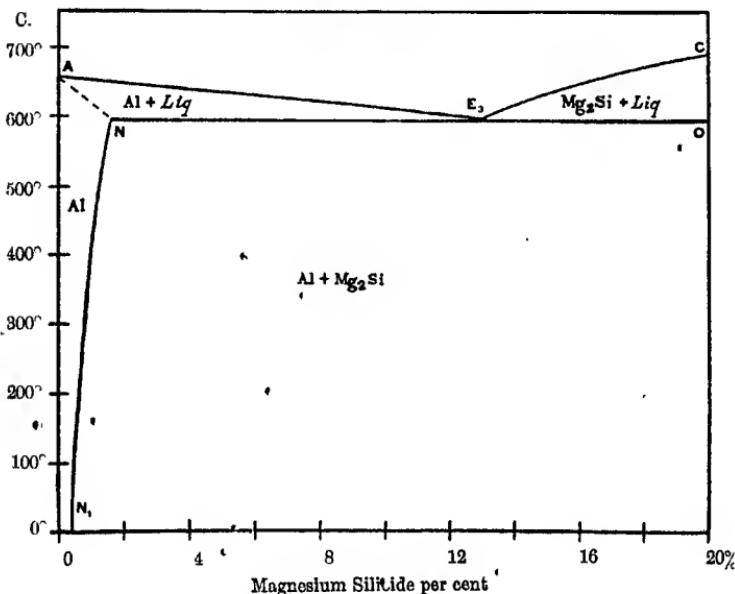
model may therefore be regarded as divided into two parts, one representing the constitution of alloys of silicon, aluminium, and Mg₂Si, and the other representing the constitution of alloys of magnesium, aluminium, and Mg₂Si. This division is very clearly brought out on the constitutional model. The plane ANE_3OC (Figs. 142, 143 and 144) divides the system into two parts, each of which is a complete ternary system in itself.

A number of experiments have been carried out with the

object of studying in detail the constitut n of this binary system Al-Mg₂Si. Thermal curves were taken with alloys whose composition corresponds to various memb⁻⁻s of this system with the object, partly of verifying its binary ch⁻racter and partly of studying the solubility of the compound Mg₂Si in aluminium. The diagram obtained from the⁻e experiments is shown in Fig. 154. The eutectic point lies at 13 per cent of the compound (8·2 per cent magnesium and 4·8 per cent silicon), while the limit of solubility at the eutectic temperature is 1·6 per cent of the compound (1·0 per cent magnesium and 0·6 per cent silicon). As the temperature falls the solubility of this compound decreases very appreciably as shown by the line NN₁ (Fig. 154). An examination of alloys which had been made homogeneous at 580° C. and were then very slowly cooled to 30° C. showed that at this lower temperature not more than 0·54 per cent of Mg₂Si could remain in solution, while it is thought that actually this figure should be less, since it is difficult to attain equilibrium in a reasonable time at low temperatures. This may be illustrated by Figs. 155 and 156, Plate 24. Fig. 155 represents an alloy containing 0·85 per cent magnesium and 0·5 per cent silicon, quenched after prolonged annealing at 581° C. It consists entirely of homogeneous solid solution in which the compound is wholly dissolved. Fig. 156, Plate 24, shows the microstructure of the same alloy, the only difference in treatment in this case being that it was very slowly cooled to 30° C. instead of being quenched. During this slow cooling a certain amount of Mg₂Si has separated from solution. The effect of the addition of excess of either magnesium or silicon on the solubility of Mg₂Si has also been determined and is indicated by the line ZRNU in Fig. 148 (page 232). This refers to the solubility at the moment of complete solidification of the alloys. The dotted line in the same figure indicates the solubility at 150° C.

V(d). The Age-Hardening of the Alloys of Aluminium with Magnesium and Silicon.

The extent to which silicon and Al₃Mg₂ are soluble in aluminium both at high and low temperatures, has been indicated above (pages

Fig. 157.—*Aluminium-Magnesium-Silicide.*
Ultimate Stress.

OUENCHED AND AGED FOR 7 DAYS
AT ROOM TEMPERATURE

SLOWLY COOLED IN FURNACE

Metric tons per sq. inch.

Magnesium Silicide por cent

Fig. 158.—*Aluminium-Magnesium-Silicide.*
Hardness Curve.

OUENCHED AND AGEO FOR 7 DAYS
AT ROOM TEMPERATURE

SLOWLY COOLED IN FURNACE

Hardness Number, Brinell.

Magnesium Silicide per cent
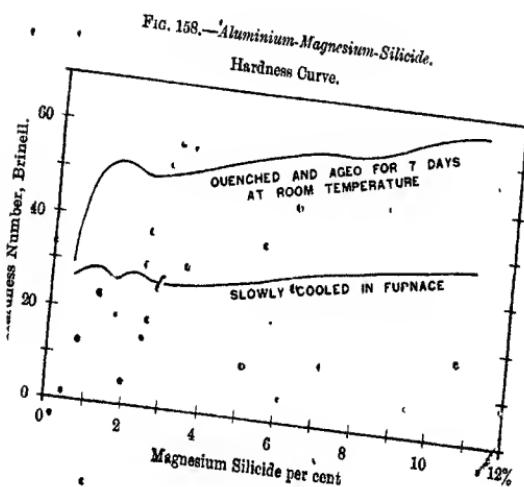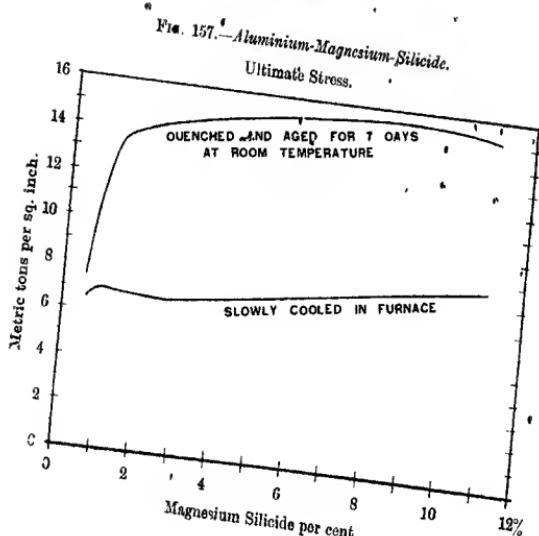
219 and 221). In the case of silicon, no variation of solubility with temperature has been found, while in the case of Al_3Mg_2, there is only a small decrease of solubility with falling temperature. On the other hand, the compound Mg_2Si is much more soluble at high temperatures than at low. In view of the fact, which has already been established, that magnesium and silicon in combination confer on aluminium the property of hardening after quenching, experiments have been

FIG. 159.—*Hardness of Quenched and Aged Alloys plotted against Composition.*
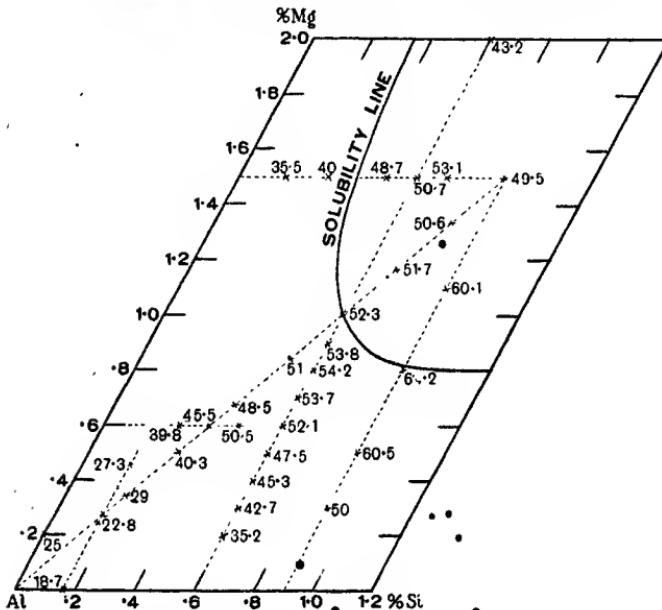
undertaken with alloys in which the compound Mg_2Si and aluminium are the only constituents, in order to ascertain the influence of Mg_2Si on the age-hardening qualities. A series of alloys has, therefore, been prepared, whose compositions are represented by points along the line NO, Figs. 143–145. In these alloys the amount of Mg_2Si increases progressively up to a concentration well beyond the limit of solubility of this compound in solid aluminium at a high temperature.

Two chill castings of each of these alloys were prepared. One was
forged hot from one inch diameter (as cast) to a thickness of $\frac{3}{4}$ inch,
and used for hardness tests ; the other was forged and rolled into a
strip $\frac{1}{10}$ inch thick and used for tensile tests. The samples were
annealed for one hour at a temperature of 500° C., after which
treatment one set of samples was quenched in water and the other

Fig. 160.—*Difference between Tensile Strengths of Aged and Slowly-Cooled
Specimens, plotted against Composition.*
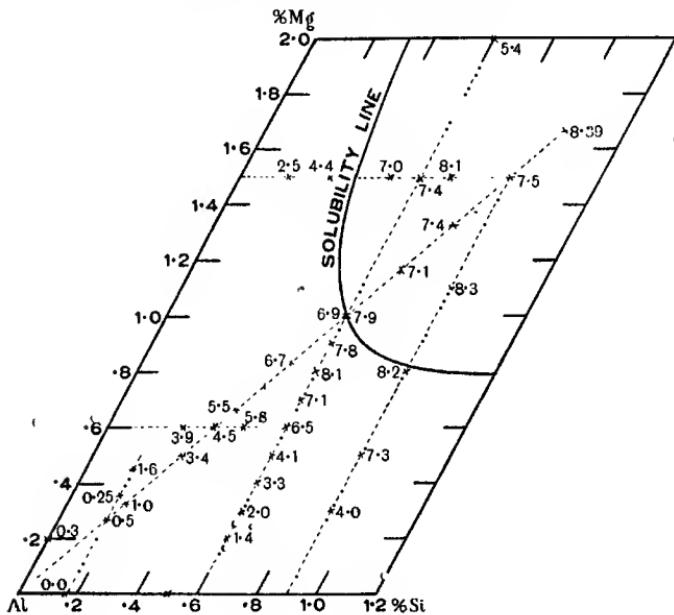
slowly cooled in the furnace. After a period of about seven days,
mechanical tests were carried out on the alloys. Results are shown
in Figs. 157 and 158 (page 236), representing tensile and "Brinell"
hardness tests respectively. The study of these diagrams shows that
the increase in hardness produced during ageing of the alloys after
quenching rises progressively with the amount of Mg_2Si present
in the alloy, until the limit of solubility is reached, beyond which

the total increase in hardness remains pra tically constant. It would therefore appear that the extent of the age-hardening which takes place is roughly proportional to the amount of magnesium silicide in solution at the moment of quenching. Further to confirm this view, other experiments were undertaken in a similar manner on other alloys of the ternary system—not necessarily belonging to the binary aluminium-Mg_2Si system—and therefore lying to either side of the line NO. The alloys examined are indicated in Figs. 159 and 160, and the results of the tests are also indicated in these Figures and in Figs. 161 and 162 (page 240). Here again in each of the series of alloys examined it is found that the hardness due to ageing increases with the amount of Mg_2Si in solution up to the limit of solubility at the quenching temperature. In Figs. 159 and 160 the limit of solubility at high temperatures is indicated by the full black line, and it will be observed that the alloys examined reach a maximum hardness in the neighbourhood of this solubility line.

The results which are summarized in these diagrams leave little doubt that the ageing of these alloys is due to the fact that, on quenching, the compound Mg_2Si is retained, probably in solution, in an unstable condition from which it tends to revert to a more stable form. It would appear that the gradual hardening is due to the changes which occur in the alloy as the result of the tendency to the formation of separate crystals of aluminium and Mg_2Si, existing side by side in the alloy. That the separation of particles of Mg_2Si of *large* size does not take place is established by microscopic examination, and further confirmed by the mechanical tests, since an alloy in this condition has a tensile strength and hardness considerably less than that of the quenched alloy immediately after quenching and before ageing takes place. An age-hardened alloy appears to reach a hard metastable state, which is either permanent, or at least persists for a period of many years. All the evidence at the Authors' disposal suggests that the alloys in the quenched and age-hardened condition may be regarded as permanent at ordinary temperatures. Some of the material produced in the present research has been under observation for periods up to ten years, while samples of commercial duralumin, first tested in

FIG. 162.

FIG. 161.

Tensile Strength,
Tons per sq. inch.

Hardness, Brinell.

Silicon per cent

Magnesium per cent
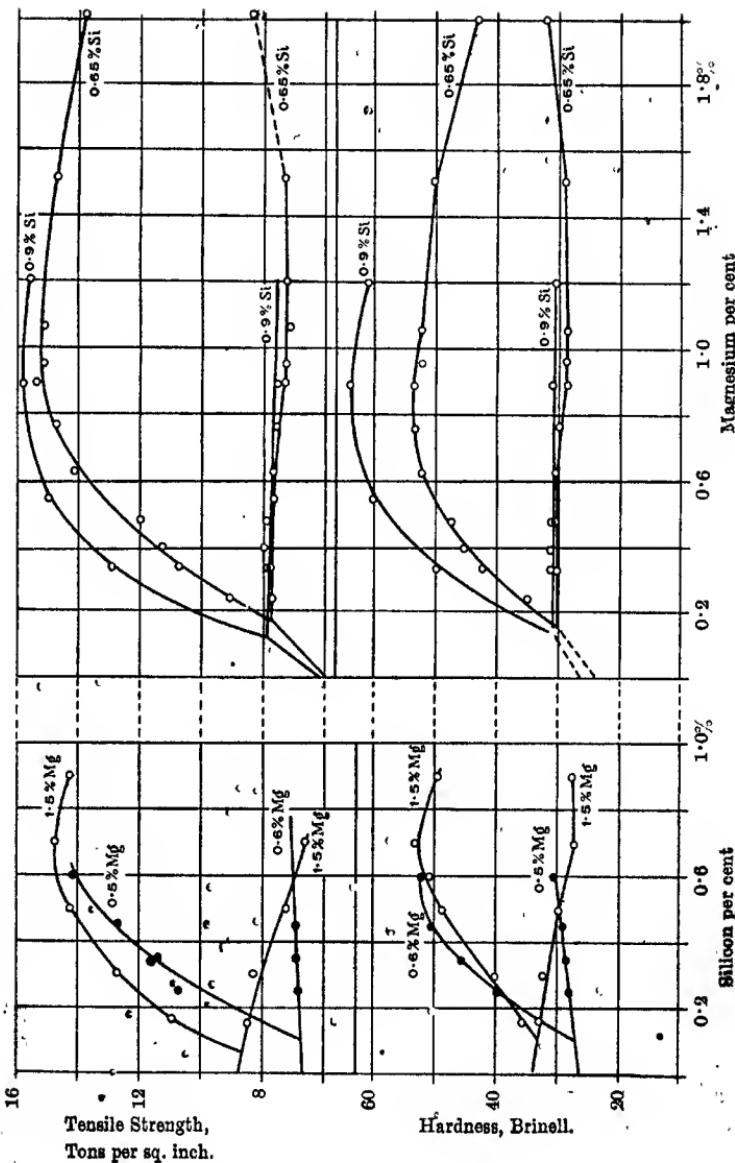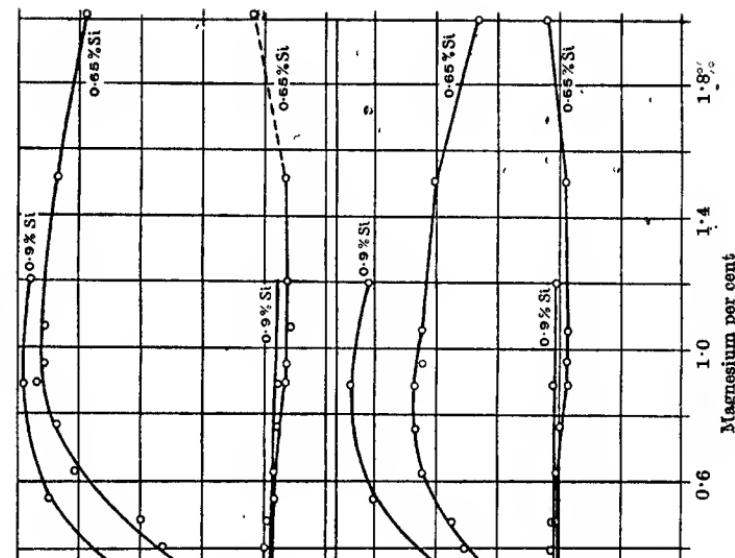
1912, have been re-tested in 1920, giving results which only indicate slight further hardening. It is possible that this condition is attained when the precipitated particles have reached a definite size (or exist within certain limits of size), and that a condition of increased hardness is associated with this particular degree of dispersion.*

RATE OF HARDENING DURING AGEING.

In the case of commercial Duralumin, it is usually found that about four days are required in order to obtain complete hardening

FIG. 163.
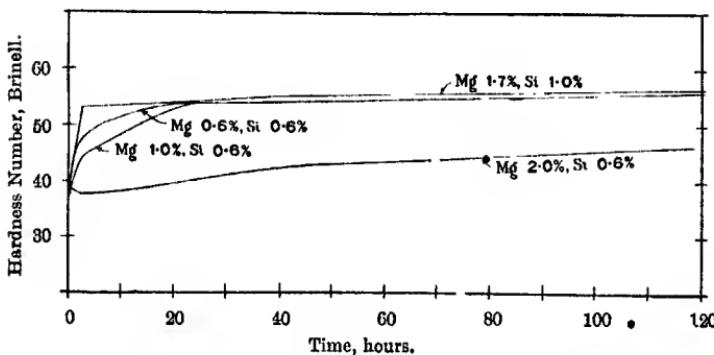
of the alloy at the ordinary temperature, although the greater part of this hardening takes place within the first twenty-four hours. Experiments were undertaken in order to ascertain the rate of hardening of the alloys now under discussion, and the curves shown in Fig. 163 indicate the rate of hardening of three alloys of the following compositions :—

1	Magnesium	. .	1	per cent.
	Silicon .	. .	0·6	,, Remainder Aluminium.
2	Magnesium	. .	0·6	,,
	Silicon .	. .	0·6	,, Remainder Aluminium.
3	Magnesium	. .	2	,,
	Silicon .	. .	0·6	,, Remainder Aluminium.

* Jeffries. Journal Institute of Metals, 1919, No. 2, page 329.

After forging, specimens of these alloys were quenched, and then tested at intervals while ageing was proceeding. It will be observed from the curves that alloys 1 and 2 harden relatively rapidly, the full hardness being attained in about four days; subsequent hardening after this time is very slow. On the other hand alloy No. 3, which contains an excess of magnesium over that required to form the compound Mg_2Si, hardens much more slowly, and the process does not appear to be complete after twelve days' ageing. Furthermore, in spite of the fact that the amount of silicon in the alloy is the same as in alloy No. 2, and that the amount of the compound Mg_2Si present is greater than that in alloy No. 2, the amount of hardening on ageing is less; the reason for this is not far to seek, since a reference to Figs. 159 and 160 shows that the solubility of this compound in solid aluminium at high temperatures is rapidly decreased by the addition of magnesium, and when the magnesium content reaches 2 per cent the amount of compound which will dissolve is only about $0\cdot7$ per cent. Considerably less total hardening would therefore be expected.

Effect of Ageing at a Temperature Higher than Atmospheric.

Since it appears that the hardening of alloys of this type is due to the fact that the quenched solid solution which exists at a high temperature cannot remain unaltered at room temperature, it appeared to be worth while to investigate the effect on this solid solution of ageing at temperatures higher than atmospheric, since it seemed probable that by suitable adjustment of conditions better mechanical properties could be obtained. The alloys used in determining the rate of hardening were therefore re-annealed at successively increasing temperatures, which are indicated in Fig. 164, the specimens being annealed for one hour at each temperature. Both hardness tests and tensile tests were made on samples after this treatment, with the results indicated in the curves of Fig. 164. At first the effect of an increase of temperature is not considerable, although it is quite marked at 100° C. With higher temperatures the mechanical properties of the alloy are improved until a maximum

is reached in the neighbourhood of 200° C. Annealing at still higher temperatures results in a rapid fall of both hardness and tensile strength in all these alloys. It is interesting to note that the same optimum

FIG. 164.

Fig. 164.

	Mg	Si
a	0·6%	0·6%
b	1·0%	0·6%
c	1·7%	1·0%
d	2·0%	0·6%
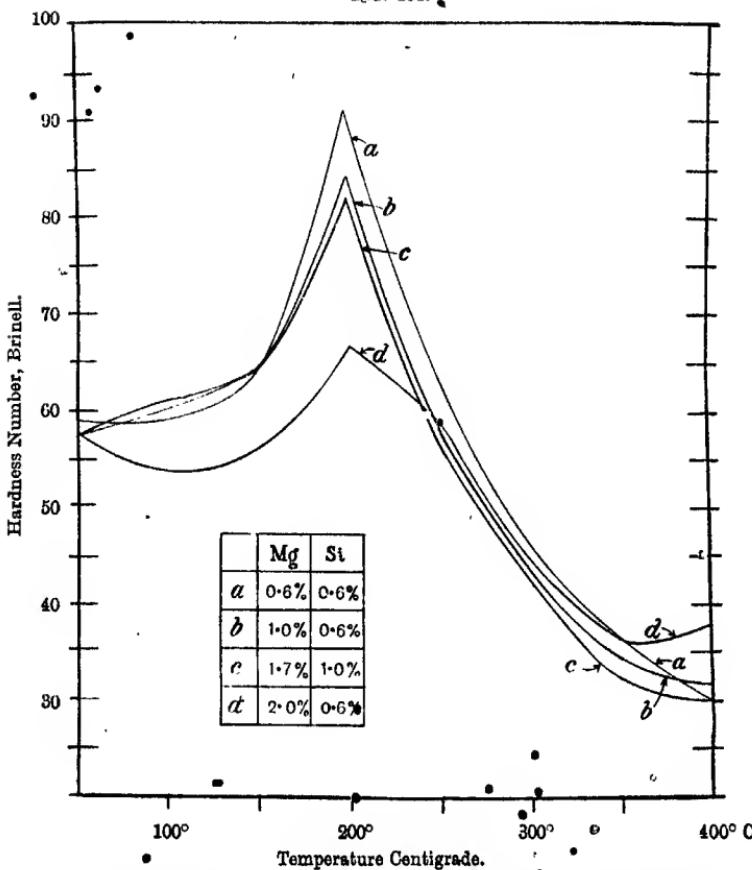
temperature is shown for all the alloys, namely 200° C.; re-heating to this temperature gives a remarkable improvement in the results of tensile and hardness tests when the alloys are compared either with the slowly cooled samples or with samples tested immediately

after quenching. The figures given show, both in the case of tensile
strength and hardness tests, an increase to about three times the
original value. There seems to be little doubt that in the case of
all the alloys of this system the best results can be obtained by re-
heating, after quenching, to a slightly elevated temperature, which,
however, may vary slightly from alloy to alloy, although the
experiments carried out suggest that heating at about 200° C. will
give the best results.

HARDENING OF SILICON-MAGNESIUM-ALUMINIUM ALLOYS
CONTAINING COPPER, ZINC OR NICKEL.

Experiments have been undertaken with the object of
ascertaining the respective effects of copper, zinc, and nickel on
the age-hardening produced in aluminium by the compound Mg_2Si.

Copper.

A constant copper content of 3 per cent has been used throughout
this series of alloys. The amount of magnesium silicide has been
progressively increased from the minimum possible, corresponding
to the silicon content of the aluminium used, to nearly 3 per cent.
Two samples of each of the alloys were annealed at 500° C.; one
sample was then quenched in water and the other cooled slowly in
the furnace. After being allowed to age for three weeks at room
temperature, hardness tests were made on these samples, and the
results are shown in the curves of the lower portion of Fig. 165.
Age-hardening in the presence of copper is quite considerable when
the amount of magnesium silicide present is 0·55 per cent
(corresponding to a silicon content of 0·2 in the alloy). The
hardening effect, however, appears to reach a maximum when the
total amount of magnesium silicide is about 0·9 per cent. (0·35 per
cent silicon). The solubility at 500° C. of magnesium silicide in
aluminium containing 3 per cent of copper is about 0·9 per cent,
corresponding to a silicon content of about 0·35 per cent. The
addition of copper to aluminium therefore reduces the amount of
Mg_2Si which can be dissolved, and in consequence reduces the
maximum hardening effect which it is possible to obtain. It was,

FIG. 165.—3% Copper Series.

EFFECT OF 1 HOURS' TEMPERING ON ALLOYS AGED AT ROOM TEMPERATURE

----- AGED 3 WEEKS

200°C.

250°C.

100°C.

50°C.

150°C.

QUENCHED AND AGED 3 WEEKS AT ROOM TEMPERATURE

SLOWLY COOLED IN FURNACE

Magnesium Silicide per cent

Hardness Number, Brinell.
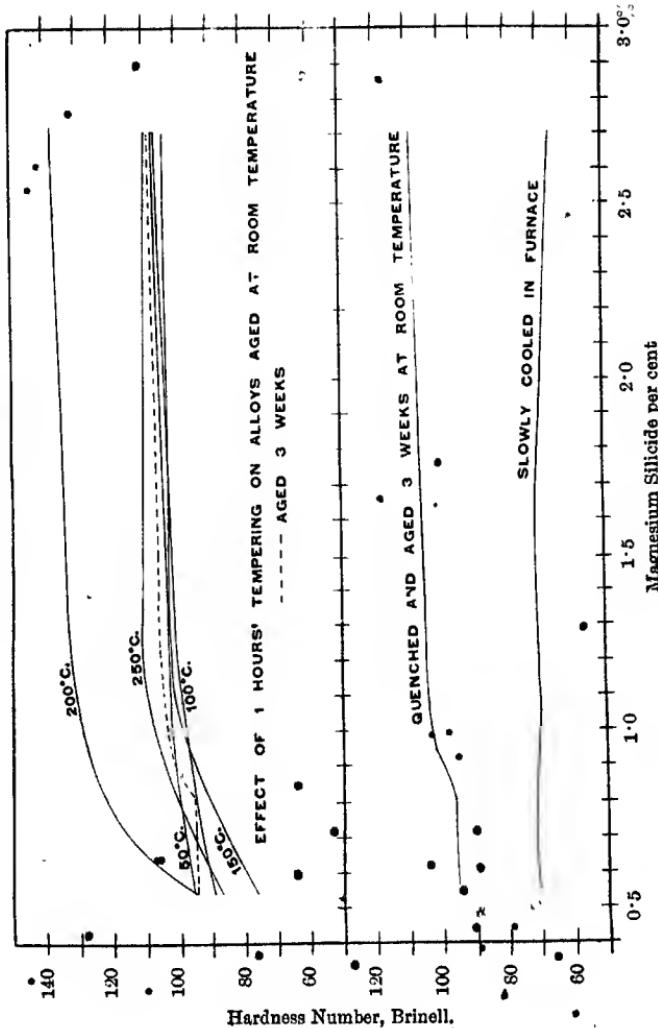
unfortunately, impossible to obtain samples of aluminium containing less than 0·2 per cent silicon, so that the effects produced by very small quantities of Mg_2Si could not be studied.

Little advantage appears to be obtained in the presence of 3 per cent of copper by increasing the content of Mg_2Si beyond about 0·9 per cent.

Experiments have been undertaken with a view to determining the effect of re-heating quenched and aged alloys to temperatures higher than that of the atmosphere. The samples used for the previous experiments were re-heated for periods of one hour at 50° C., 100° C., 150° C., 200° C., and 250° C. Hardness tests were carried out after each treatment, and the results are plotted in the curves shown in the upper portion of Fig. 165. It is seen that re-heating to any temperature up to 100° C. has but little effect on the alloys. At 150° C, alloys containing very low amounts of Mg_2Si become somewhat reduced in hardness ; the other variations in hardness observed within this range of treatment are within the errors of experiment. Heating at 200° C. results in a marked increase in the hardness of all the alloys except that containing the lowest amount of Mg_2Si. Re-heating to 250° C. causes the hardness to fall again. The general character of these results is somewhat similar to those obtained with the pure ternary alloys of aluminium, magnesium, and silicon.

Zinc.

Similar experiments have been carried out on alloys containing 15 per cent of zinc instead of 3 per cent of copper. The corresponding curves are shown in Fig. 166, and are very similar to those of the copper alloys. In this case, however, it is noticed that the improvement on tempering is very slight, and takes place at a lower temperature, namely, 150° C. Re-heating to 200° C. causes a very marked fall in the hardness of the alloys.

Nickel.

In these experiments 3 per cent of nickel has been used. Nickel appears to have less effect on the solubility of Mg_2Si in aluminium

FIG. 166.—15% *Zinc Series.*

EFFECT OF 1 HOURS' TEMPERING ON ALLOYS AGED AT ROOM TEMPERATURE
----- AGED 3 WEEKS

150°C.
50°C.
100°C.
200°C.

QUENCHED AND AGED 3 WEEKS AT ROOM TEMPERATURE

SLOWLY COOLED IN FURNACE

Magnesium Silicide per cent

Hardness Number, Brinell.
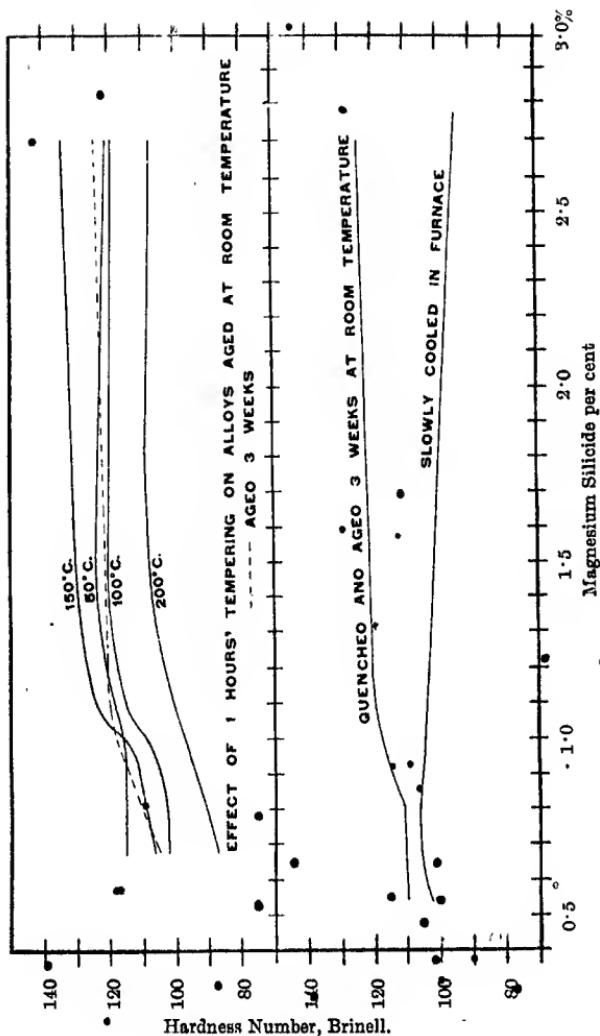
than has either copper or zinc. The curves for this series, Fig. 167, show a very close resemblance to those found in the pure ternary alloys ; the maximum hardening effect is found with a magnesium silicide content of about 1·3 per cent, and on re-heating 'the hardened alloys at temperatures up to 200° C. the hardness is steadily increased. Re-heating to 250° C. results in a very marked fall in the hardness.

In general it may be said that the addition to aluminium alloys containing small quantities of magnesium and silicon of other metals which are frequently found in commercial light alloys, results in a diminution of the solubility of the compound Mg_2Si, and to some extent modifies the hardening effect which can be obtained. At the same time the properties of these alloys can be improved in a marked degree by controlling the composition and the heat treatment. For example, in the case of a 3 per cent copper alloy containing 1 per cent of magnesium silicide, the hardness after slow cooling is about 70, by quenching and ageing at room temperature this hardness can be increased to over 100, while further heat treatment at a temperature of 200° C. raises the hardness still further to about 130, or nearly double the hardness figure of the slowly cooled alloy. Similarly in the case of an alloy containing 3 per cent of nickel together with 1·3 per cent of magnesium silicide, the hardness can be increased from 35 in the slowly cooled specimen to 70 in the specimen quenched and aged at room temperature, and to nearly 100 by suitable further treatment at 200° C.

In the case of the zinc alloys, the improvement obtained, while quite appreciable, is distinctly less than in either of the other series.

Ageing of Copper-Aluminium Alloys.

It has recently been suggested* 'that the hardening of "Duralumin" and similar alloys is due to the retention in solution,

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* Merica, Waltenberg and Scott. American Inst. Mining and Metallurgy. June 1919, page 913.

Fig. 167.—8% *Nickel Series.*

EFFECT OF 1 HOURS' TEMPERING ON ALLOYS AGED AT ROOM TEMPERATURE

------ AGED 7½ DAYS

200°C.

150°C.

100°C.

250°C.

QUENCHED AND AGED 7½ DAYS AT ROOM TEMPERATURE

SLOWLY COOLED IN FURNACE

Magnesium Silicide per cent.

Hardness Number, Brinell.
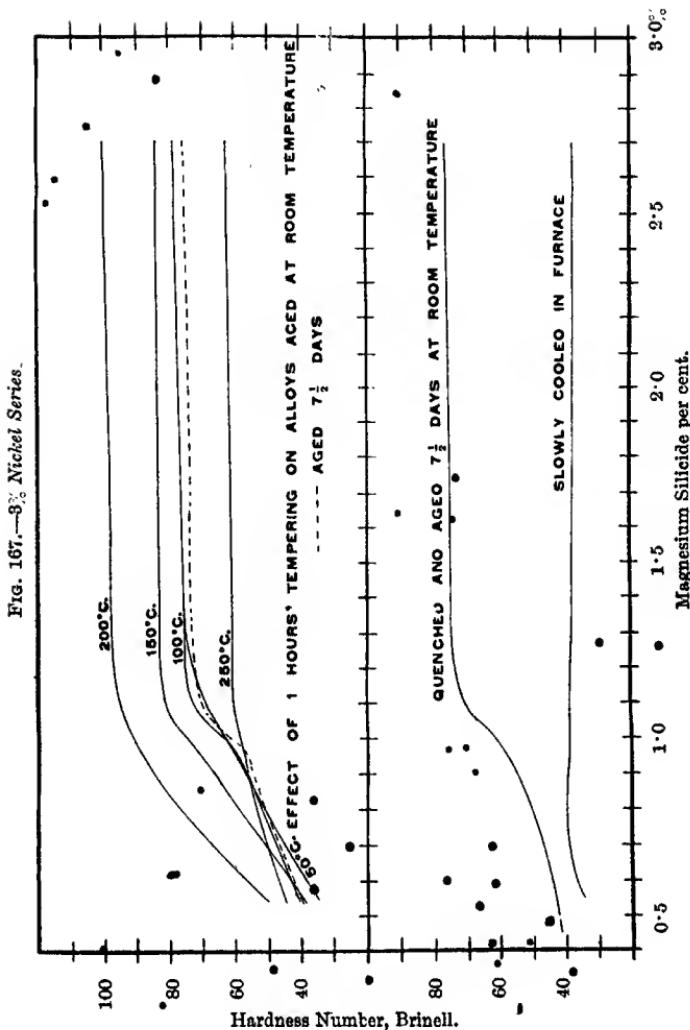
as a result of quenching, of the compound $CuAl_2$, and that the partial re-precipitation of this compound at room temperature is responsible for the hardening produced.

FIG. 168.—*Copper-Aluminium Series.*
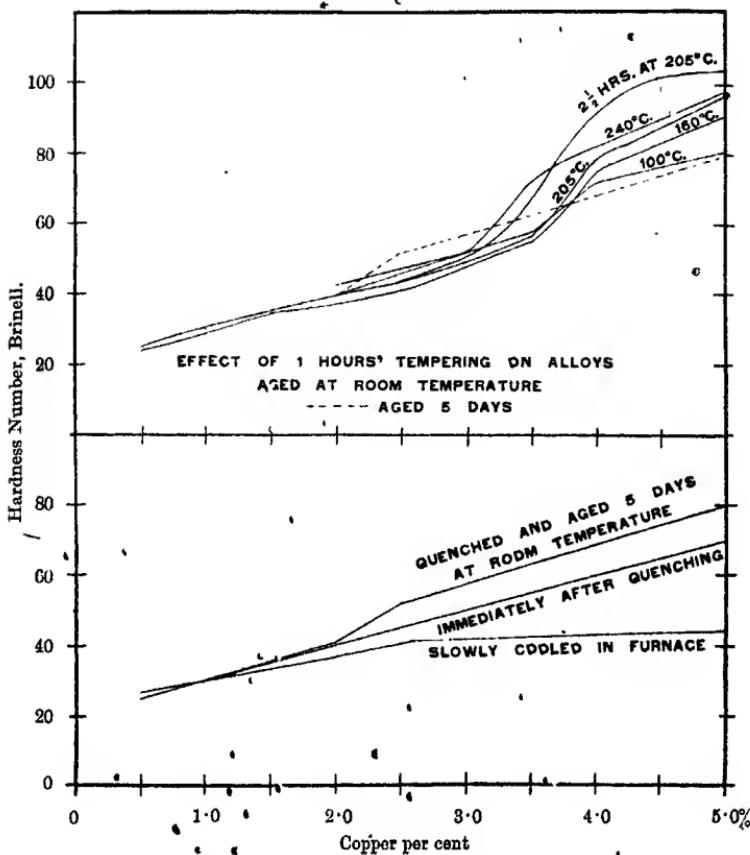
Experiments have been undertaken with the object of studying the hardening of copper aluminium alloys after quenching from a high temperature. Ten alloys containing 0·5 to 5 per cent copper have been cast, forged, and annealed at 500° C. After this

treatment one sample of each alloy was cooled in the furnace, while another was quenched in water. The quenched alloy was tested immediately after quenching, and again at successive intervals up to five days. The hardness of these alloys after different treatments is shown in Fig. 168. In the lower part of the Figure the hardness of the slowly cooled alloys is compared with that of the quenched alloys both immediately after quenching and after five days ageing. An ageing effect becomes apparent in all alloys containing from 2·5 per cent to 5 per cent of copper, but the actual increase in hardness is only small. In no case was an increase of hardness greater than 10 in the Brinell figure obtained. This effect is quite insufficient by itself to account for the hardening produced in "Duralumin."

These alloys were re-heated to successively higher temperatures for a period of one hour and the results of this treatment are shown in the upper portion of Fig. 168. The effect of re-heating at 100° C. is slight; there appears to be, in some of the alloys, a definite softening as the result of this re-heating treatment, and this softening effect continues until an annealing temperature of about 200° C. is reached, except in the case of alloys containing more than 3·5 per cent copper, where a slight increase in the hardness is noticed. Annealing for a longer period at 205° C. increases the hardening effect observed; a higher temperature still, namely 240° C., causes a marked softening in those alloys containing more than 3·5 per cent copper. It would appear from these results that the maximum hardness is obtained with an alloy containing about 4·5 per cent of copper, which has been quenched from 500° C. and subsequently re-heated for about three hours at 205° C. The amount of this hardening, however, is considerably less than that found in alloys containing magnesium silicide; in some cases suitable treatment of alloys containing this latter compound has resulted in a trebling of the Brinell hardness number.

A diagram recently published in America * indicates a maximum

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* Merica, Waltenberg and Freeman. Sci. Paper, No. 337. U.S. Bureau of Standards (1919).

solubility of about 4 per cent of $CuAl_2$ at 500° C. and 1 per cent at room temperature. The determinations carried out in the course of the present research, on the contrary, show a solubility of nearly 5 per cent at 500° C. and about 3 per cent at room temperature. The difference between these results probably arises from the fact that the methods used by the American investigators were not adequate to secure complete equilibrium at the lower annealing

FIG. 169.
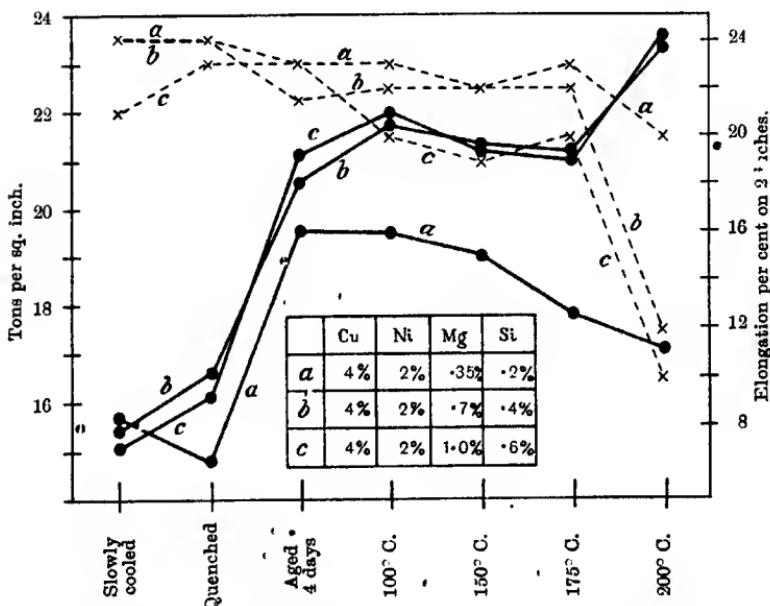
temperatures, since enormous times are required to obtain equilibrium in these conditions by annealing alone. The method adopted at the National Physical Laboratory consists in heating the alloys to a temperature near the solidus for a prolonged period to obtain the maximum solution of the compound in the solid aluminium; this is followed by extremely slow cooling to lower temperatures, from which the alloys are quenched. The present Authors consider that, provided the rate of cooling is extremely slow, this method will

give more reliable results than simple annealing at very low temperatures. It is of interest to note that the curves shown on Fig. 168 (page 250), confirm the view that the solubility of $CuAl_2$ in aluminium at room temperature is much greater than 1 per cent,

Fig. 170.

	Cu	Mg	Si
a	4%	·35%	·2%
b	4%	·85%	·5%
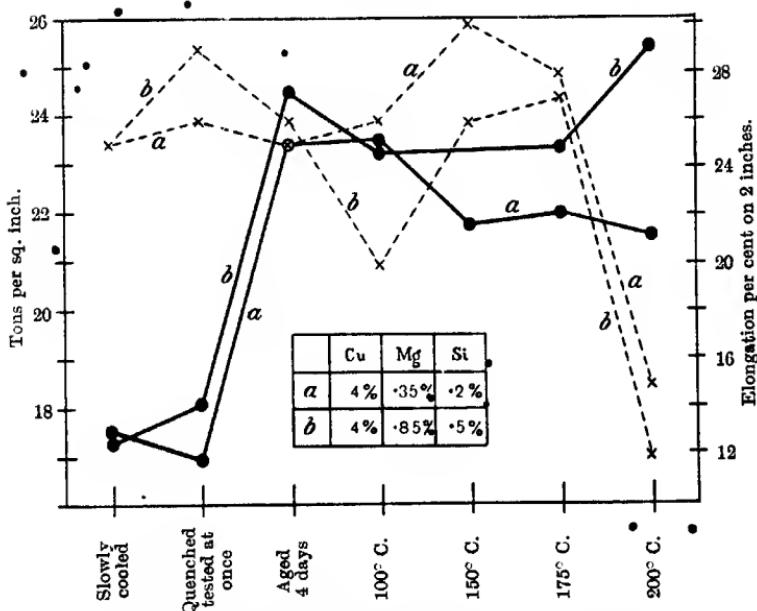
since no increase in hardness on ageing is observed in alloys containing less than 2·5 per cent of copper.

HEAT-TREATMENT OF ALUMINIUM ALLOYS CONTAINING COPPER, NICKEL AND MANGANESE TOGETHER WITH MAGNESIUM SILICIDE.

Experiments have been undertaken with a view to determining the strengths of certain alloys in which the content of magnesium silicide was carefully regulated. Three series of alloys have been used for these experiments ; the first contained 4 per cent of copper and 2 per cent of nickel, and was of the same type as the " Y "

alloy described elsewhere in this Report. The second series contained
4 per cent of copper, while the third contained 4 per cent of copper
plus 0·5 per cent manganese. These last alloys approach closely
to the composition of commercial Duralumin. The amounts of
magnesium silicide added to these alloys are shown in Table 55.
All the alloys were cast in metal moulds, annealed, forged, and rolled
into strip $\frac{1}{16}$ of an inch thick. Samples of each alloy were then
submitted to the following treatments :—

Sample A.—Annealed at 500° C., slowly cooled in furnace.

Sample B.—Annealed at 500° C., quenched in water.

Sample C.—Annealed at 500° C., quenched in water, aged 4 days
　　at room temperature.

Sample D.—Annealed at 500° C., quenched in water, aged 1 hour
　　at 100° C.

Sample E.—Annealed at 500° C., quenched in water, aged 1 hour
　　at 150° C.

Sample F.—Annealed at 500° C., quenched in water, aged 1 hour
　　at 175° C.

Sample G.—Annealed at 500° C., quenched in water, aged 1 hour
　　at 200° C.

TABLE 55.

Mark.	Copper.	Nickel.	Manganese.	Magnesium.	Silicon.
	Per cent.	Per cent.	Per cent.	Per cent.	Per cent.
V 1402	4	2	—	0·35	0·2
V 1403	4	2	—	0·7	0 4
V 1404	4	2	—	1·0	0·6
V 1405	4	—	—	0·35	0·2
V 1406	4	—	—	0·85	0·5
V 1407	4	—	0·5	0·85	0·5
V 1408	4	—	0·5	0·85	0·2

Results of mechanical tests on these alloys are shown in Figs. 169, 170 and 171. Fig. 169 (page 252), which refers to the alloys containing 4 per cent of copper and 2 per cent of nickel, shows that with a low content of magnesium silicide (0·55 per cent), the tensile

Fig. 171.

	Cu	Mn	Mg	Si
a	4 %	·5%	·35%	·2 %
b	4 %	·5 %	·85%	·5 %
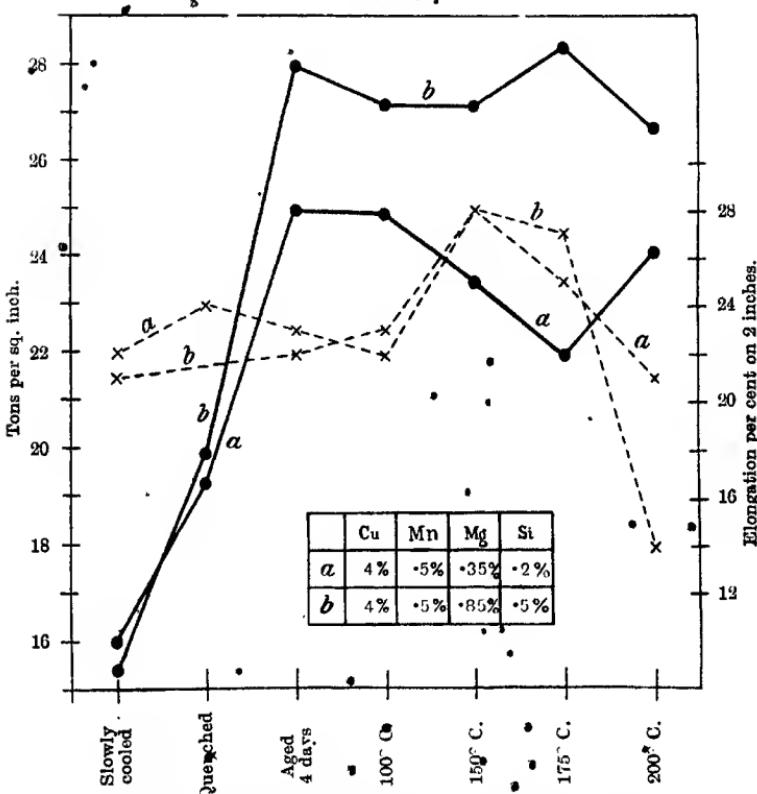
strength developed after hardening and ageing is not very high, and that further treatment at higher temperatures results in a reduction in the strength of the alloy. An increase in the amount of magnesium silicide to 1·1 per cent gives after quenching and ageing a stronger alloy, which is capable of improvement by further re-heating; a

further increase in the amount of magnesium silicide does not appear to confer any improved properties.

In the case of the alloy containing 4 per cent of copper (Fig. 170), the strength of the alloy aged at room temperature does not appear to vary greatly with a higher magnesium silicide content; the stability of the alloy at higher temperatures, however, is decidedly improved by the addition of magnesium silicide.

In the case of the alloy containing 4 per cent of copper with 0·5 per cent of manganese (Fig. 171), the amount of the magnesium silicide appears to be of considerable importance. An improvement of three tons per square inch in the tensile strength is found in the alloy aged at room temperature, as a result of increasing the content of magnesium silicide from 0·55 per cent to 1·35 per cent; furthermore, the superior strength of the alloy containing the higher amount of Mg_2Si is retained after re-heating to temperatures up to 200° C.

The elongation of these alloys is indicated by the dotted lines; it will be noticed that the increase of strength due to age-hardening is not accompanied by a reduction in the elongation; re-heating the age-hardened alloys to higher temperatures does not lower the elongation appreciably until a temperature of 200° C. is reached, but after heating at this temperature the value is notably reduced in all the alloys used.

The Report is illustrated by Plates 1 to 24 and 92 Figs. in the letterpress.

Fig. 1. *Examples of Alloy " A " in various stages, including Forged Billet Extruded and Rolled Rods, thick Blank and finished Sheet.*
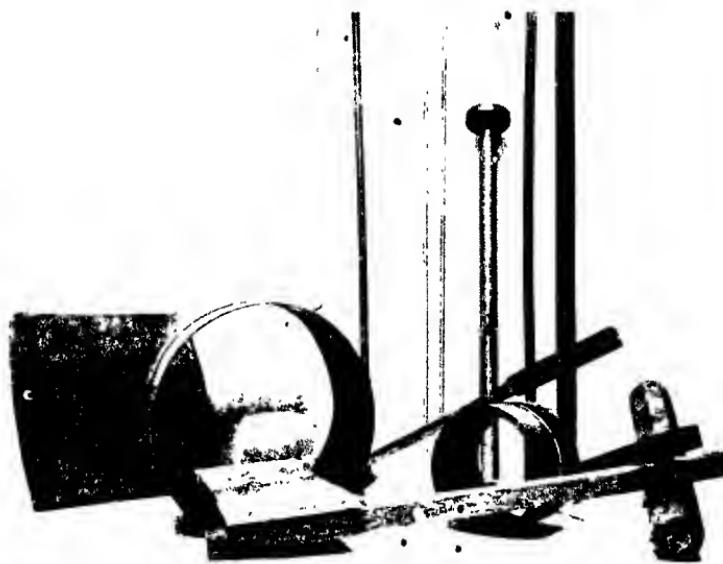
Fig. 2. *Girders in "A" and "E" Alloys.*
Very thin Sheet, Spinnings, Pressings, Wire, in "A" Alloy.
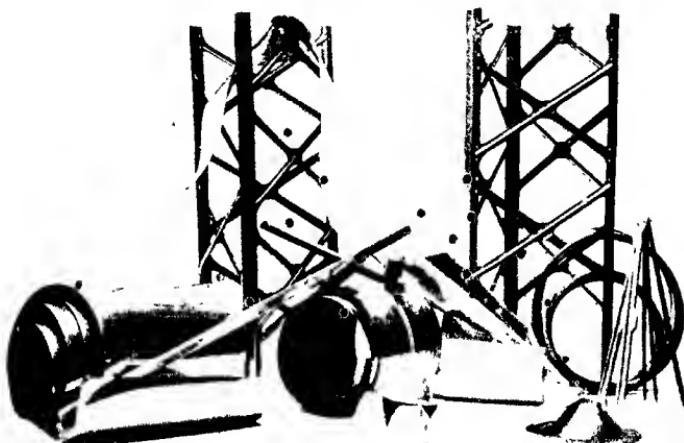
Figs. 7 and 8.   *Copper-Zinc-Aluminium—Copper 3%, Zinc 20%.*
     Sand Cast.   *Chill Cast.*
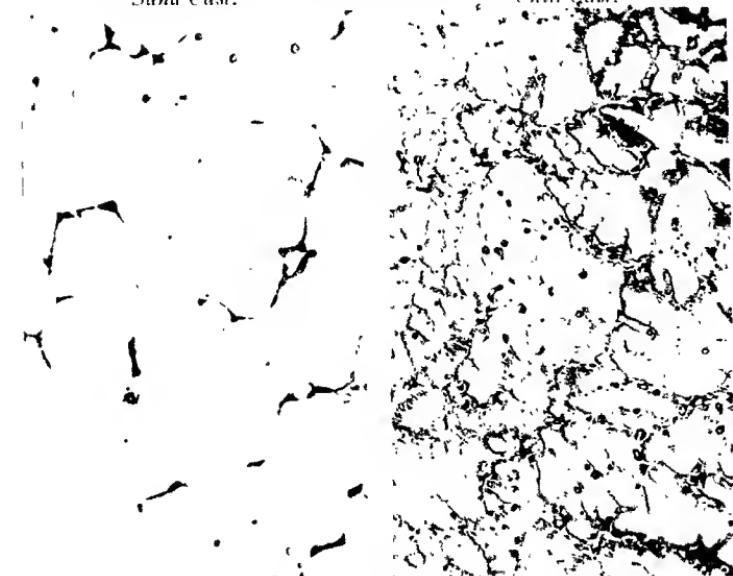
Figs. 9 and 10.   *Copper 1%, Zinc 20%.*
  Sand Cast.   *Chill Cast.*
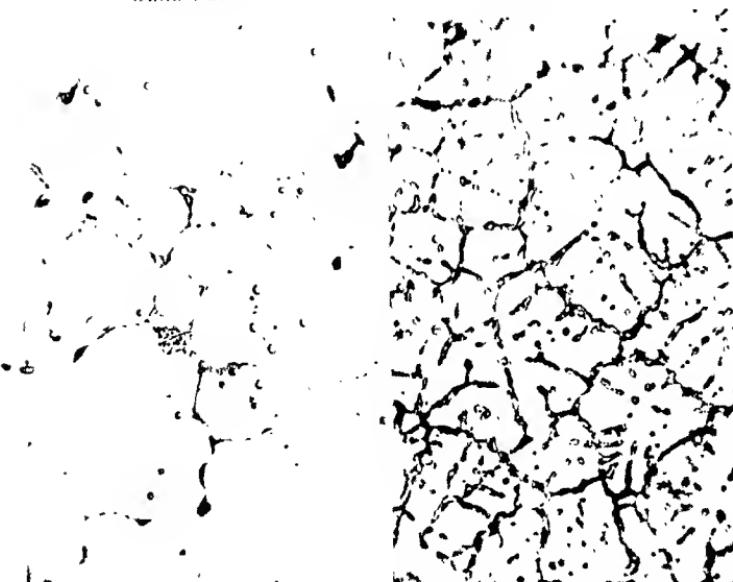
Fig. 23. As Cast. Unetched.
× 150 d.

Fig. 24. As Cast. Etched H.F
(Ni Al, darkened.) ×400 d.
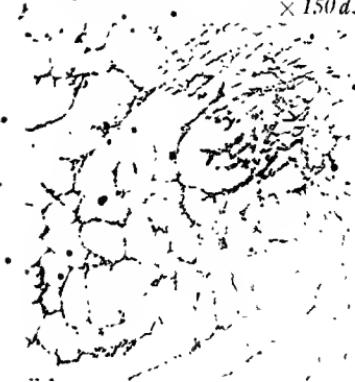
Fig. 25. As Cast. Etched hot.
HNO₃. (Cu Al₂ darkened.) × 400 d.

Fig. 26. Annealed 6 hours 530°C
Quenched cold water. Unetched.
× 150 d

Fig. 27. Same as Fig. 26.
× 500 d.

Fig. 28. Same as Fig. 26, Re-heated
to 500°C and Slowly Cooled. ×500 d

Plate 4.   **ALLOYS RESEARCH.**
Rolled Rod · 150 diams. (*See page* 66.)

Fig. 41.   *Copper 1%, Zinc 5%.*   Fig. 42.   *Copper 3%, Zinc 5%.*

Fig. 43.   *Copper 1%, Zinc 20%.*   Fig. 44.   *Copper 3%, Zinc 20%.*

Mechanical Engineers 1921.

Fig. 45. (p. 68.) *Experimental Foundry at the National Physical Laboratory*
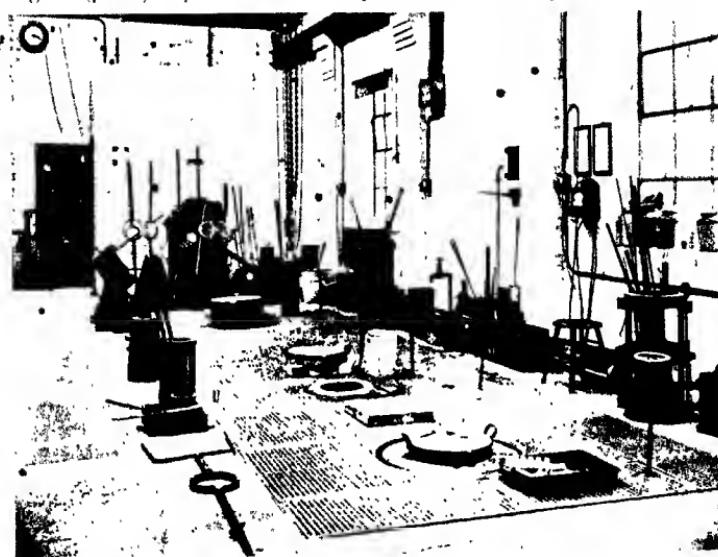
Fig. 46. (p. 69.) *Machine for combined clamping and tilting of Slab Moulds.*

Mechanical Engineers 1921.

Plate 6.   ALLOYS RESEARCH.
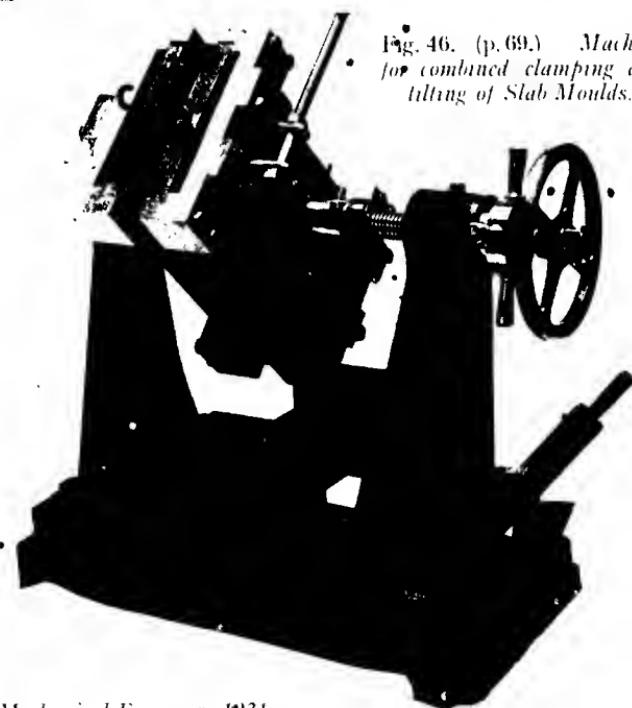
Fig. 47. (p 69) *Experimental Rolling-Mill at the National Physical Laboratory.*

Fig. 48. (p. 70) *Gas-burners for Pre-heating Experimental Rolls.*
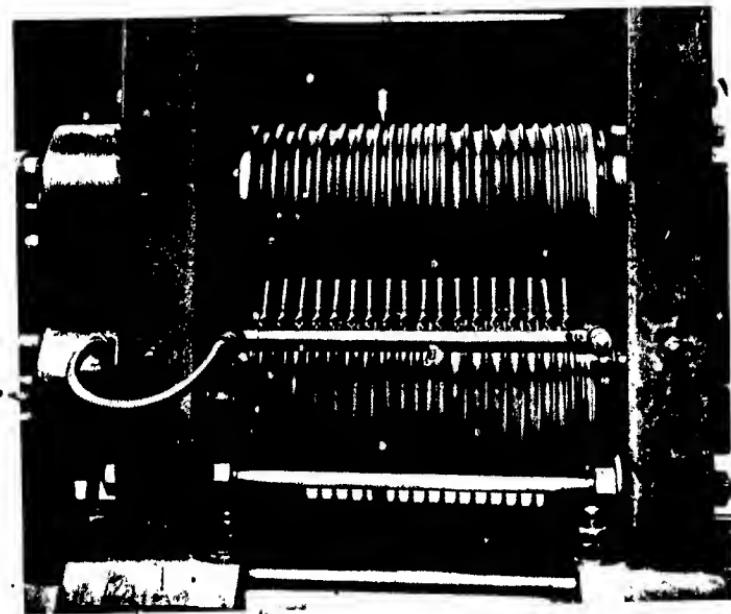
Fig. 49. (p. 70.)   *Electrically-Heated Muffle, 12 in. diameter,*
                          for Heating Light Alloys.
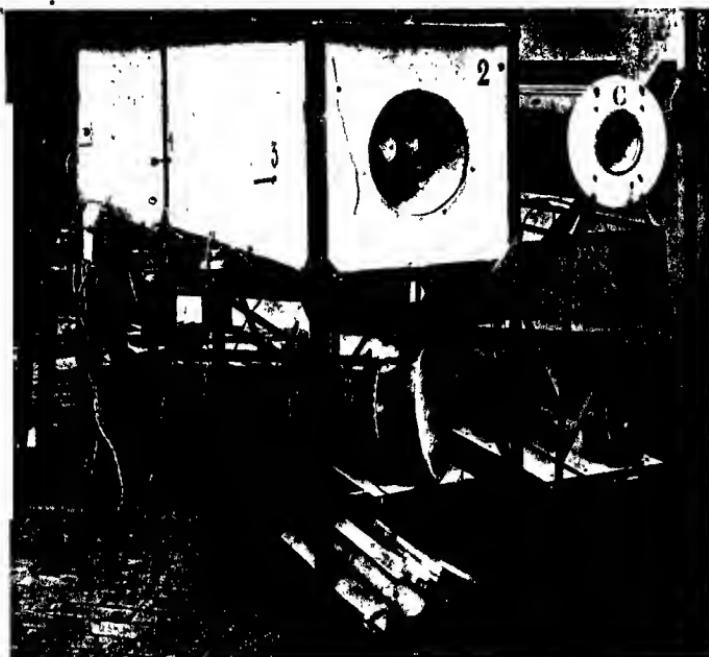
Fig. 50. (p. 70.) *Larger view of separately Heated Stopper*
for securing uniform temperature in the above Muffle.

Mechanical Engineers 1921.

Plate 8.   ALLOYS RESEARCH.
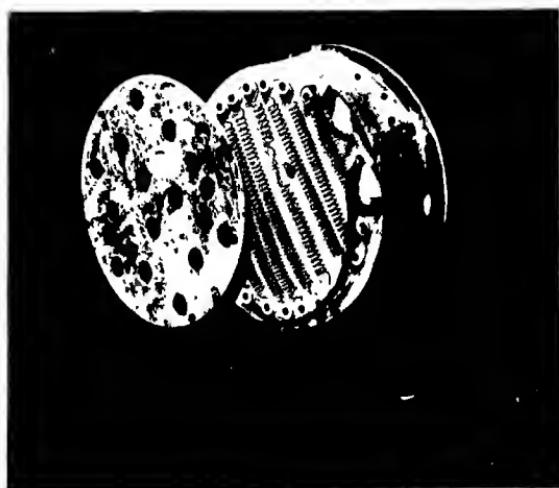
Alloy "A." (See page 75)

Fig. 53.   Twice Forged.
Fig. 52.   As Cast.   150 d.   Annealed 400°C.   ×. 150 d.
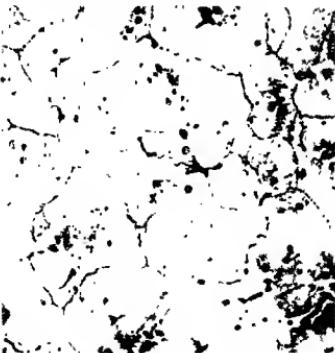
Fig. 54.   Twice Forged. Annealed 400°C.   Re-annealed
500°–520°C. (Alcoholic soda etch in all cases.)   × 150 d.

Fig. 59. (p. 87.)   Holes in very thin sheet made by rifle bullets—front and
back views.
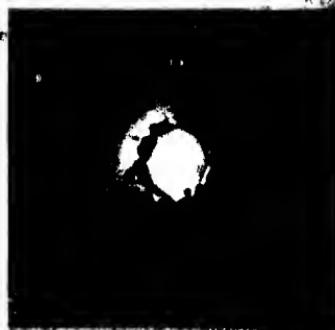
Fig. 67. Forged 10%
Unetched. × 45 d.

Fig. 68. Same as Fig. 67.
Unetched. × 500 d.

Fig. 69. Forged 10% Annealed
480°C. 30 mins. Quenched.
Etched H.F. × 150 d.

Fig. 70. 0·75 Blank. Annealed
18 hours 480°C. Quenched.
Unetched. × 150 d.

Plate 10. ALLOYS RESEARCH.
(See page 129.)

Fig. 71. Alloy " Y." 0·2 in. Plate,
Air-cooled after Rolling.
Unetched. 150 d.

Fig. 72. Alloy " Y." Finished
Sheet, Hot-rolled and Heat-treated.
Etched. 500 d.

Fig. 73. Alloy " Y." Finished
Sheet. Cold-rolled and Heat-
treated. Etched. × 500 d.
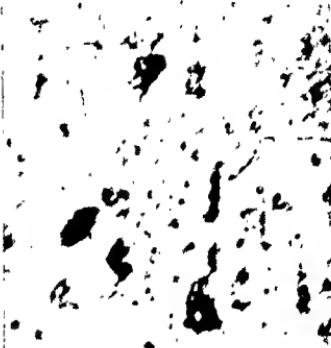
Fig. 74. Alloy " W. 847."
Magnesium 1·8%, Nickel 3%,
Copper 6%. Hot-rolled. . 150 d.

Fig. 75. Alloy " W. 871." Magnesium 1%, Nickel 3%,
Copper 6%. Chill Cast. × 150 d.

Fig. 77.   Alloy "A." Extruded
1¼ in. round Rod   150 d.

Fig. 76.
Extruded Section 5 in.   1¼ in.
"E" Alloy
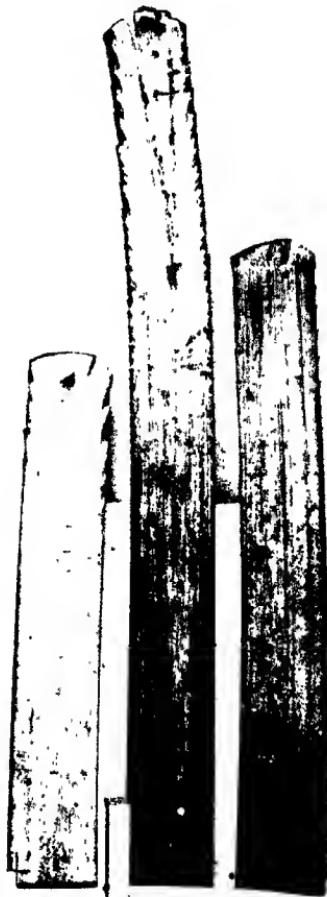
Fig. 78.   Alloy "A." Extruded
2¼ in. round Rod   150 d.

Plate 12.   ALLOYS RESEARCH.

Fig. 79. Alloy "A." Extruded 1⅛ in. round Rod, showing corrugated crystal boundaries,   × 150 d.
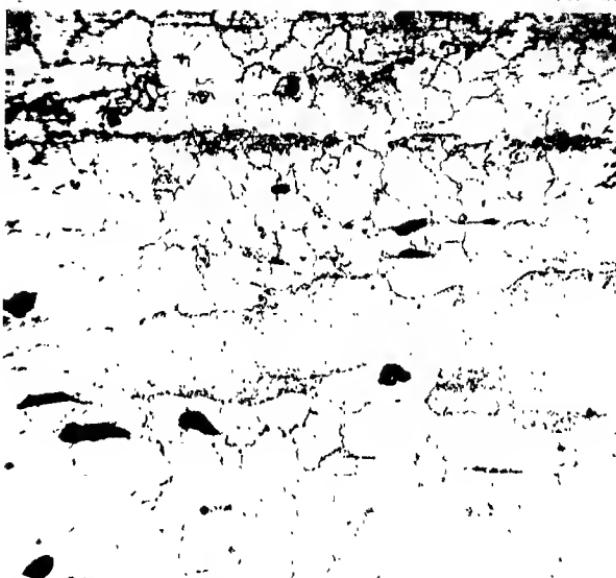
Fig. 80. Alloy "A." Extruded Rod, showing large elongated areas.   × 150 d.
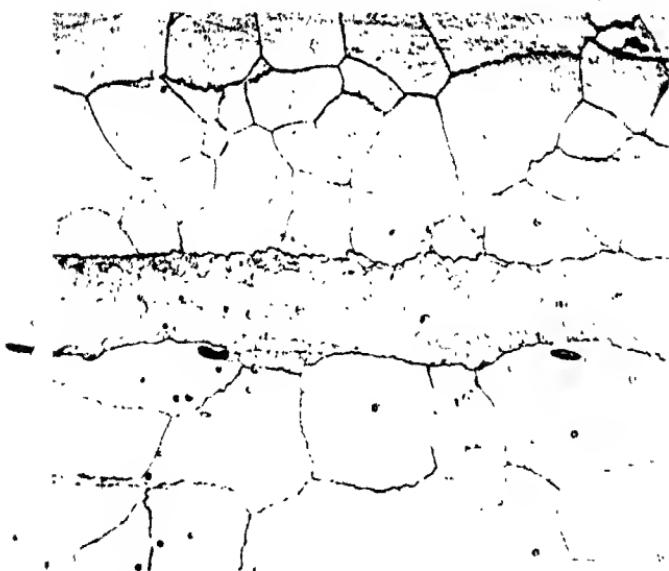
(See page 141)

Fig. 81. *Alloy "A."   Extruded Rod, showing effect of annealing
on elongated areas*   × *150 d.*
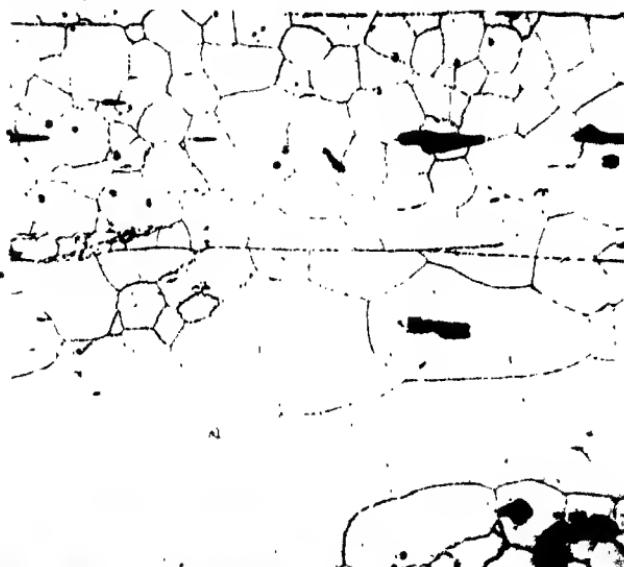
Fig. 82. *Alloy "A." Extruded Rod, showing serious lamination
of the core.*   × *150 d.*

Plate 14. ALLOYS RESEARCH.
(See page 341).

Fig. 83. *Alloy "B (3/25)." Extruded Rod, showing presence of dark-etching constituent.* 600 d.

Fig. 84. *Alloy "E." Extruded and rolled. Etched 25% HNO_3.* 150 d.

Fig. 85. *Alloy "E." Extruded and rolled Etched KOH* 150 d.

× 150 diams.

Fig. 92. (p. 180.) *Alloy "A." Inter-crystalline crack in over-annealed structure.*

Fig. 93. (p. 181.) *Alloy "A." Annealed at 250°C., immune from inter-crystalline cracking.*

Fig 94. (p 185.) *Alloy "A"+0.5% Manganese Annealed 6 hours at 450 C.*

Fig. 95. (p.188.) *Alloy "E." Quenched from 400°C.*

Fig. 96. (p. 191.) *Corrosion Test-pieces in Teak Frame after prolonged immersion in the sea. (Aluminium-Zinc Alloys.)*

Plate 16. ALLOYS RESEARCH.
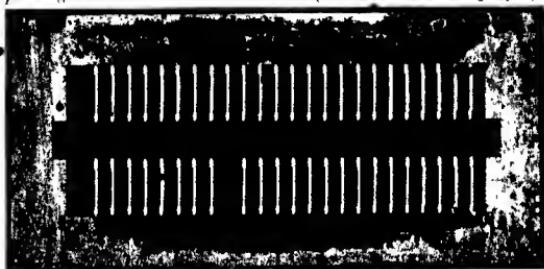
Ill Etched with 10% NaOH in Water. 150 diams. (See page 209.)

Fig. 108. *Copper 10%, Zinc 40%, as Cast.*

Fig. 109. *Copper 20%, Zinc 30%, as Cast.*

Fig. 110. *Copper 25%, Zinc 30%, as Cast.*

Fig. 111. *Copper 20%, Zinc 45%, as Cast.*

Mechanical Engineers 1921.

ALLOYS RESEARCH.   Plate 17
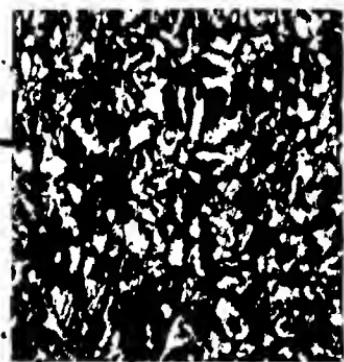
Ill Etched with 10% NaOH in Water.   > 100 diams. (See pages 209 and 210.)

Fig. 112. *Copper 20%, Zinc 30%, as Cast.*

Fig. 113. *Same as Fig. 112. Re-heated to 360°C., and Quenched.*

Fig. 114. *Same as Fig. 112. Annealed 20 hours at 405°C., and Quenched.*

Fig. 115. *Same as Fig. 112. Re-heated to 425°C., and Quenched showing signs of melting.*

Plate 18. ALLOYS RESEARCH.

All Etched with 10% NaOH in Water (See page 270)

Fig. 116. *Same as Fig. 112. Re-heated to 440°C., and Quenched, showing further melting.* × 400 d.
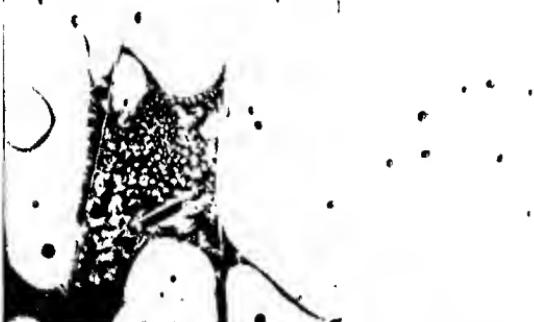
Fig. 117. *Copper 10%, Zinc 20%. Annealed 420°C. Slowly Cooled to 220°C., and Quenched.* 150 d.

Fig. 118. *Copper 10%, Zinc 25%. Treatment as in Fig. 117.* × 150 d.
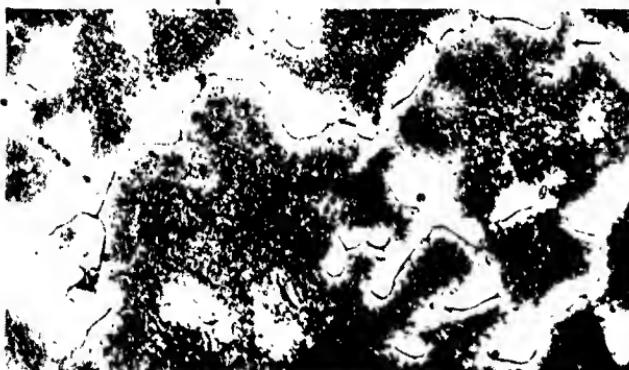
Fig. 119a (p. 210.) *Copper 26°₀, Zinc 5.°₀. Annealed 420°C.*
Slowly Cooled to 220°C, and Quenched
Etched 10°₀ NaOH in Water.   150 d.
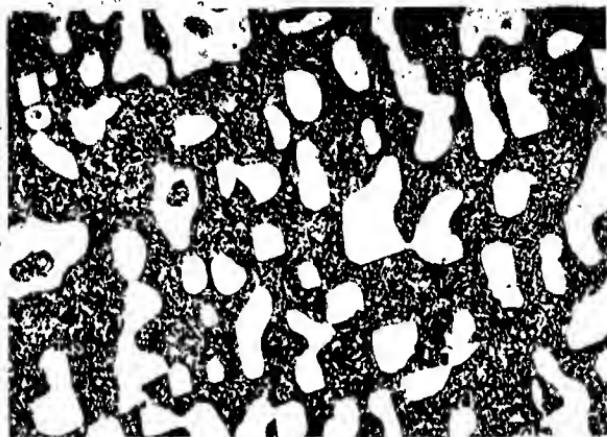
Fig. 101. (p. 204.) *Constitutional Model. Copper-Zinc-Aluminium Alloys.*

Plate 20.   **ALLOYS RESEARCH.**
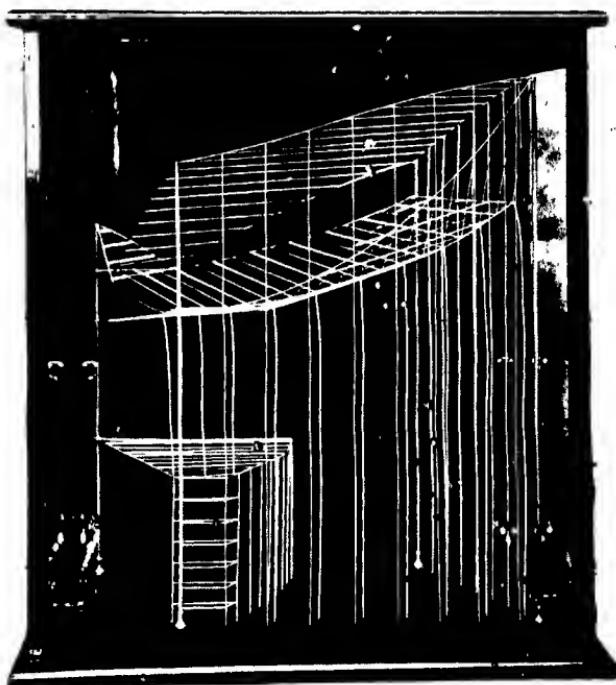
All Etched with 10% NaOH in Water. × 1,000 diams. (See page 416.)

Fig. 126.   Silicon 1%, Iron 8%.   Annealed at 565°C.

Fig. 127.   Silicon 2%, Iron 8%.   Annealed at 565°C.
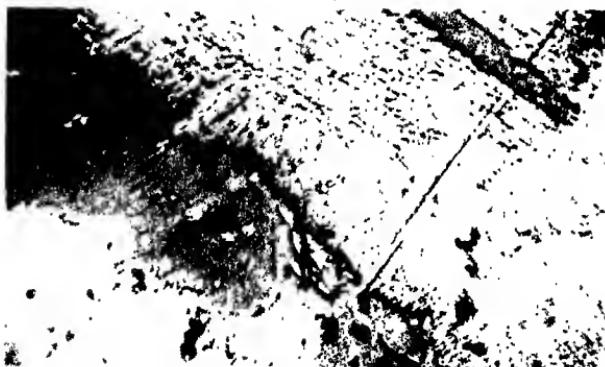
Fig. 128.   Silicon 8%, Iron 1%.   Annealed at 565°C.
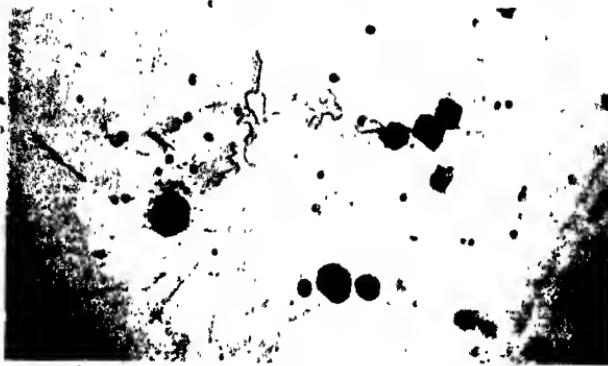
Aluminium Magnesium Alloy etched with 10% HNO; in Alcohol. × 150 diams. *(See pages 219 and 220.)*

Fig. 131. *Magnesium* 10%. *An-nealed 420°C. Slowly Cooled to 160°C., and Quenched.*

Fig. 132. *Magnesium 20%, as Cast.*

Fig. 133. *Magnesium 35%, as Cast.*
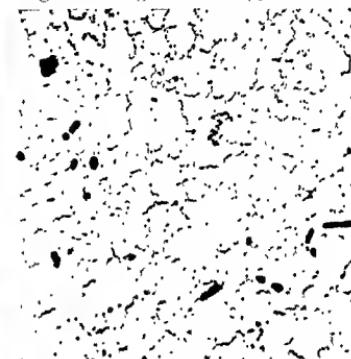
Fig. 134. *Magnesium 90%, as Cast.*

Fig. 137. (p. 222.) *Aluminium-Silicon Alloy, 3% Silicon, as Cast.* × 150 d.

Mechanical Engineers 1921.

Plate 22.   ALLOYS RESEARCH.

Fig.141. (p. 227.) *Constitutional Model. Silicon -Magnesium–Aluminium Alloys.*
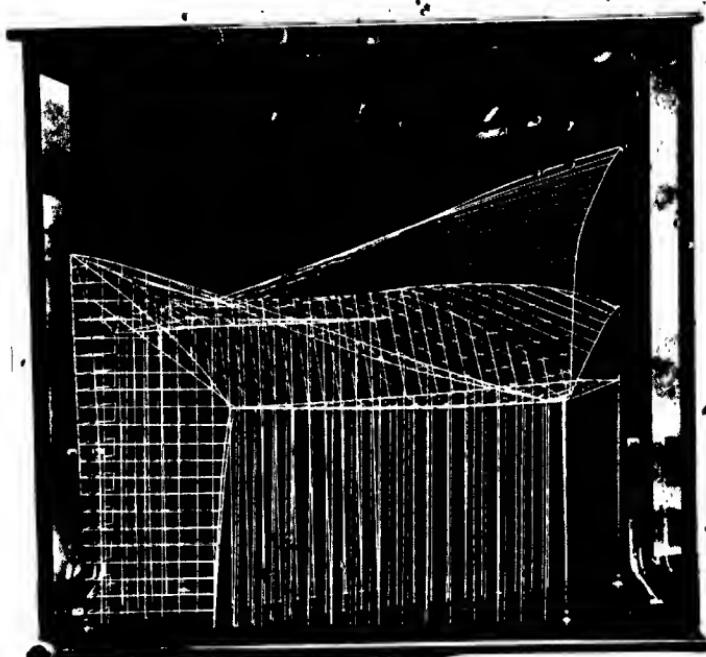
Fig. 149. (p. 231.)   *Magnesium* 5%, *Silicon* 4%, *as Cast.*
Unetched.   × *150 d.*

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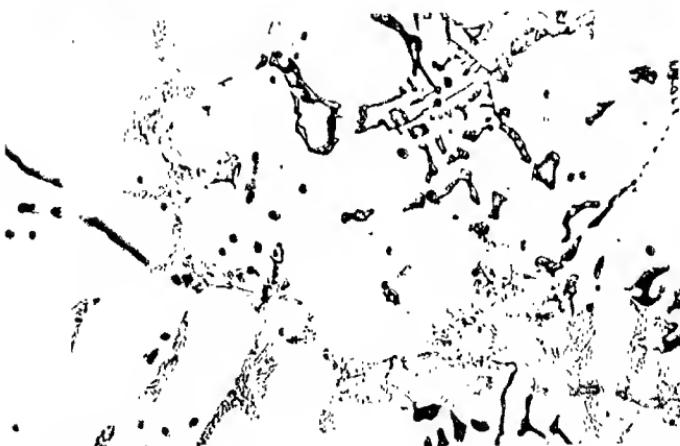
· Fig. 150.   *Magnesium 10%, Silicon 4%, as Cast.   Unetched.*
× 150 d.

Fig. 151.   *Magnesium 10%, Silicon 8%, as Cast.   Unetched.*
· 150 d.

Fig 152.   *Magnesium 25%, Silicon 2%, as Cast.*
Etched with 10% HNO₃ in Alcohol.   × 200 d.
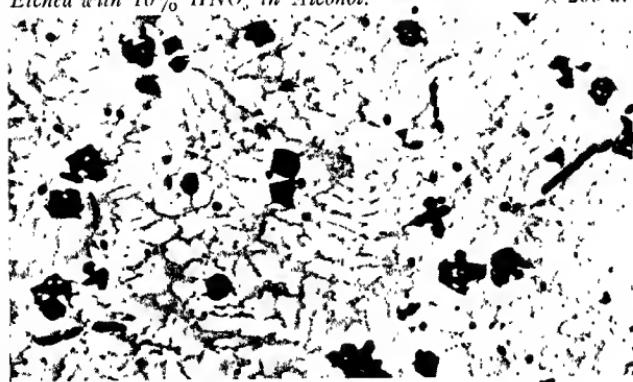
Fig. 153.   *Magnesium 35%, Silicon 10%, as Cast.   Unetched.*
× 150 d.
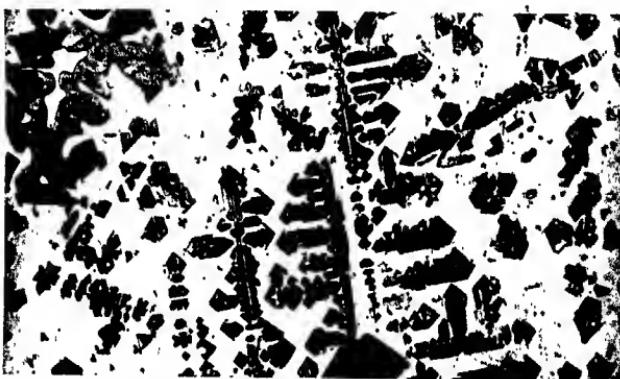
Fig. 155.   *Magnesium 0·85%,   Silicon 0·5%.   Annealed at
581°C., and Quenched.   Unetched*   × 1,000d.

Fig. 156.   *Magnesium 0·85%,   Silicon 0·5%.   Annealed at
581°C., and Slowly Cooled to 30°C.   Unetched.*   × 1,000 d.